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A TEST METHOD FOR THE EVALUATION OF AQUEOUS QUENCHING MEDIA FOR STEEL HARDENING

BY F. W. TREMBOUR AND HOWARD SCOTT

Abstract

Aqueous media are the most active quenching agents available, but suffer from the disadvantage of allowing the random formation of soft spots particularly on the surface of shallow hardening steels. To study the relative merits of different aqueous solutions, a test method was developed which is based on the quantitative measurement of the area of soft spots, the formation of which has been favored by elevating the bath temperatures. The measure of the quenching efficiency of a particular solution is taken as the bath temperature at which a certain small proportion of soft spots arises with a standard specimen and procedure.

The optimum concentrations for aqueous solutions of NaCl and NaOH were determined as 9 and 3 per cent, respectively. It is also shown that additions approaching saturation of these and other salts are detrimental to quenching action. Some solutes are extremely detrimental in even very small concentrations. A slight decrease of depth of hardening was observed to accompany mounting solute concentration.

Soft spots were observed in detail and found to be of two types which differed in appearance and mode of formation, and one of which, the "island" type, affected the depth of hardening.

PLAIN carbon steels are the least expensive, are most easily machined of all the structural steels, and accordingly are the most economical to use provided other characteristics permit. The major factor antagonistic to their use in applications where high strength is required is the fact that a very fast quench is necessary for hardening. Because of this requirement complicated sections are easily distorted in heat treatment; nevertheless plain carbon steels are used quite efficiently, for example, in shafts and couplings where

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the penetration of hardening is shallow and, being symmetrical, the degree of distortion is tolerable.

There remains, however, a serious difficulty which prevents full realization of the economies inherent in the use of carbon steels. This difficulty is the random occurrence of incompletely hardened areas or soft spots on the work after quenching, these often being more prevalent at one time than at another. In an effort to determine the factors responsible for such erratic hardening, a test method was developed with which quantitative data on this elusive phenomenon were acquired.

The general problem of steel quenching has been treated by Scott,^{1, 2} who pointed out that soft spots are not found in commercial steels hardened in the submerged water spray apparatus. Also, specimens showing nonuniformity of hardness as quenched in media other than the submerged water spray have been rehardened and soft spots have reappeared, but in different places. These facts show clearly that the source of trouble is in the quenching medium and not in the steel but, of course, some steels are more susceptible to nonuniform hardening than others.

Obviously a spray quench would be a solution to the shop problem were it permissible from the standpoint of practicability. In certain cases, such as in the heat treatment of simple short cylinders, it can be applied quite effectively. It is doubtful, however, that it can be applied to most commercial shapes without causing excessive distortion, because the shape must be strictly symmetrical, the spray must be uniform around the axis of symmetry, and the work immersed exactly along this axis.

Since a spray quench is usually impracticable the ideal quench is a still liquid which will harden the surface of the work completely when the latter is immersed without agitation. Water, under favorable conditions, is the most active quenching liquid known, producing cooling rates very close to the highest theoretically attainable.¹ The problem is then not to find a better medium, but to learn how to modify and use unagitated water so as to eliminate soft spots. Soft spots being the major criterion of performance of aqueous quenching baths, the test method developed was based on their production and evaluation under controlled test conditions.

¹H. Scott, "Some Problems of Quenching Steel Cylinders," *TRANSACTIONS, American Society for Metals*, Vol. 22, 1934, p. 68.

²H. Scott, "The Problem of Quenching Media for the Hardening of Steel," *TRANSACTIONS, American Society for Metals*, Vol. 22, 1934, p. 577.

The total area of soft spots on the surface of quenched specimens of uniform size is a good index of the efficiency of the quench if soft areas actually form. They are easily revealed by a simple technique described presently. With small specimens suitable for laboratory use, however, the amount of soft area is very small or negligible with quenching into most aqueous media at room temperature. To meet the foregoing situation, tests are made with the bath

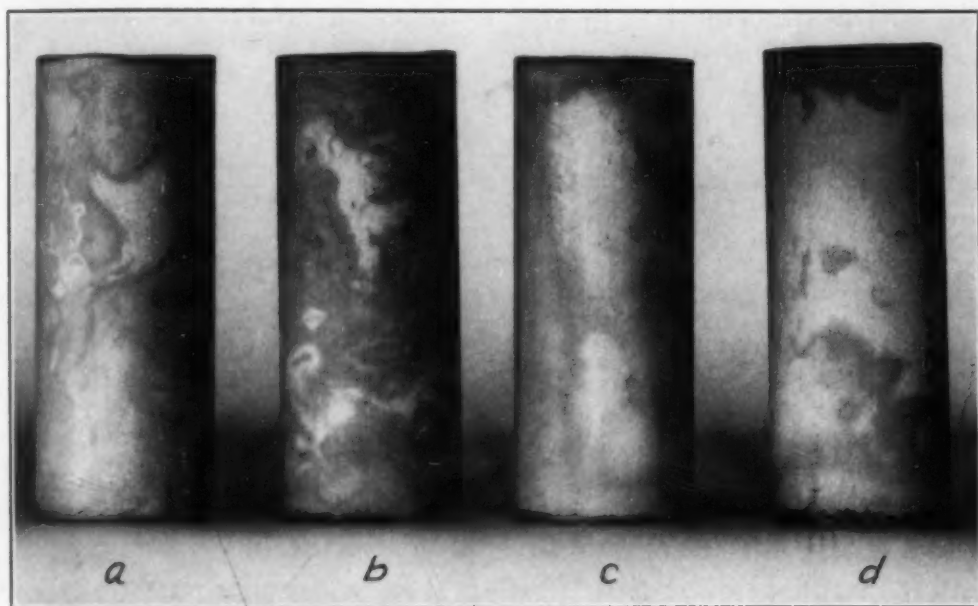


Fig. 1—Soft Spots Revealed by Etching Surface of Steel Cylinders as Quenched in Tap Water From 815 Degrees Cent. (1500 Degrees Fahr.):

Water Temperature		Soft Area
(a)	30 degrees Cent. (86 degrees Fahr.)	6.4 per cent of total
(b)	40 degrees Cent. (104 degrees Fahr.)	8.8 per cent of total
(c)	50 degrees Cent. (122 degrees Fahr.)	22 per cent of total
(d)	60 degrees Cent. (140 degrees Fahr.)	55 per cent of total

first at room temperature and then at several successively higher temperatures. The soft spot area increases rapidly with bath temperature as shown in Fig. 1. Quenching baths are then rated by a temperature at which soft spots first arise or at which a certain specified proportion of soft surface is found. This manner of rating quenching baths is quite satisfactory because it is indicative of the behavior of the tested media when quenching large and massive pieces of metal. The larger the piece, the more the temperature of the adjacent zone of quenching liquid is raised, and accordingly the medium providing freedom from soft spots at the highest bath temperature is the best for shop use so far as obtaining uniformity of hardness is concerned.

TEST SPECIMEN AND METHOD

Cylindrical test pieces of $1\frac{3}{16}$ inch in diameter and 3 inches long are used. They are of a convenient size for both the heat treating operations and for subsequent surface measurements because soft spots, when formed, are of sufficient area to be representative of the action of the medium under test, but are generally not large enough to make area determinations tedious. The test specimens used in this work were centerless ground to $1\frac{3}{16}$ inch in diameter from $1\frac{1}{4}$ -inch diameter, hot-rolled bars.

It is necessary that all pieces of a test series are in the same state of structure and of the same surface finish prior to heat treatment. The annealed or normalized structural condition is satisfactory. The surface, after removal of sufficient material to assure freedom from such defects as decarburization, may be either centerless ground, finely machined, or smoothly sandpapered. A sand-blasted, rusted, or similarly roughened surface is not acceptable since soft spots which then develop are poorly defined and difficult to measure.

In order that the sensitivity of the test be sufficiently great to permit the evaluation of the relative quenching powers of even the most effective media, a steel of high critical cooling rate was used, namely, a fine-grained S.A.E. 1045 steel which hardens to 50 per cent martensite at a depth of about 0.160 inch below the surface in $1\frac{3}{16}$ -inch diameter cylinders when quenched in submerged water spray from 1500 degrees Fahr. (815 degrees Cent.). The chemical composition of this particular steel was:

	Per Cent		Per Cent
C	0.46	Si	0.20
Mn	0.76	Cr	0.017
P	0.012	Ni	0.008
S	0.028		

Soft spots form on this steel even on quenching in unagitated water at 0 degree Cent. On the other hand, deep hardening steels develop surface soft spots only when quenched in the poorest of media at moderate bath temperatures and are therefore not suitable material for these test purposes.

It is desirable to maintain the composition of the gaseous phase surrounding the steel in the heating furnace as constant as possible during an experimental run on a series of specimens. An electrical resistance furnace, containing only common air, gives constant and

Table I

Consistency of Determinations of Soft Areas in Fine Grained S.A.E. 1045 Steel Cylinders Quenched from 816° C. (1500° F.) in Tap Water at 30° C. (86° F.).

Position of Cylinder in Quenching Bath	Soft area in per cent of total area, Specimen No.:					Deviation from mean
	I	II	III	IV	Mean	
Horizontal	6.5	6.9	6.3	6.4	6.5	- 0.2 to + 0.4
Vertical	8.0	7.5	10.9	8.7	8.8	- 1.3 to + 2.1

reproducible conditions in this respect. A furnace of sufficient heat capacity to bring the test specimen, when placed cold into the hot furnace, to furnace temperature in less than 15 minutes was used. A further 20-minute "soaking" time has been allowed before quenching, and it has been found that the total time of 35 minutes in the furnace, using a temperature of 1500 degrees Fahr. (815 degrees Cent.) for this S.A.E. 1045 steel, is not long enough to cause serious scaling or appreciable decarburization. The specimens were supported in the furnace in a horizontal position on steel racks, making contact with the racks only at the extreme ends.

Volume of the quenching medium, the quenching power of which is to be determined, is a matter of some concern. It is desirable that during quenching the temperature of the bath shall not rise materially, yet its volume should be as small as possible for ease in handling. With a 5-gallon bath, the test specimens used, when quenched from 1500 degrees Fahr. (815 degrees Cent.), raised the bath temperature only 2 degrees Cent., a sufficiently close approach to constancy. Using a quenching vessel of this capacity, 11 inches in diameter, the bath was 12 inches deep. For reasons already given, quenching was done with the least possible agitation. This state is reproducible, and, furthermore, is necessary to produce soft spots at moderate bath temperatures, a condition required to distinguish between the action of various media. Presumably the relative efficiencies of certain quenching media established by testing in the still state will retain that relationship in the agitated state.

A factor which influences the proportion of soft spots formed is the position in which the test pieces are plunged into the quenching bath. It was found that quenching the steel cylinder of the standard size adopted in a vertical position will tend to give a greater percentage of soft area, but also a greater variability in results than will quenching with the longitudinal axis of the cylinder in the horizontal position. The preferred method, therefore, and the one

used in all subsequent tests, was the latter. In this position a greater area of soft spots is found on the upper surface than on the lower. Experimental results showing the effect of position in the quench on the amount of soft spots are given in Table I.

To provide free access of the quenching liquid to all parts of the hot steel surface, light, pincer-type tongs were used, which span the steel cylinder across the ends and present minimum obstruction to the action of the cooling liquid. Another factor controlled was depth of immersion. This was conveniently done by fastening a short spur transversely to the tong handles near the joining rivet so that in the quenching operation this spur can rest against a stationary cross-piece laid on top of the quenching vessel. In this position the specimen was 4 inches below the surface of the liquid.

It was observed in the course of these experiments that the scale formed on the surface of the steel during the heating stage was invariably cast off by all of the aqueous quenching media tested. The scale was eliminated explosively upon immersion and none was found to adhere to the specimens upon withdrawal after completion of the quench. No indication of interference by scale with quenching action of the liquids was observed.

EVALUATION OF SOFT AREA

Upon removal of the test pieces from the quench the surface is dark and tarnished but quite smooth and scale-free so no special surface preparation is required before etching to reveal soft spots. The etching agent used is a 50 per cent solution of commercial concentrated hydrochloric acid in water maintained at a temperature of 85 to 90 degrees Cent. during etching. The test pieces are stood vertically on end to permit uniform acid attack on the curved surface, and an immersion period of 5 to 10 minutes at the given temperature is sufficient to develop the desired contrast.

Soft spots revealed by the etching have a light gray color as compared to a dark gray for hardened parts, when the surface is viewed normally. The characteristic appearance of the etched specimens is reproduced in Figs. 1 and 2. When, however, the surfaces are observed obliquely, at a large angle from the normal, the colorings described above will be reversed, the soft spots appearing darker. In either case the distinction is readily made and may be checked by a few confirmatory hardness readings. Soft spots will generally

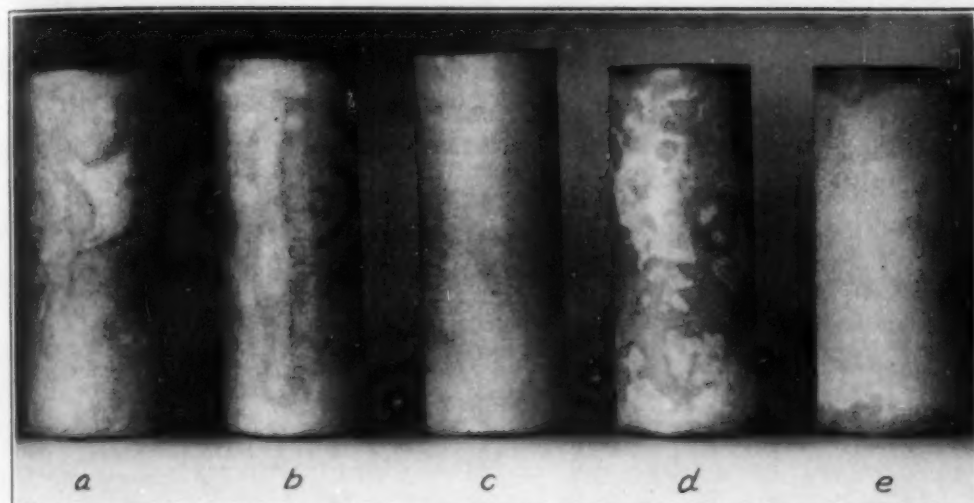


Fig. 2—Soft Spots Revealed by Etching Surface of Steel Cylinders as Quenched in Brines of Following Concentrations at 30 Degrees Cent. (86 Degrees Fahr.) From 815 Degrees Cent. (1500 Degrees Fahr.):

Brine Concentration		Soft Area
(a)	0 per cent sodium chloride	6.4 per cent of total
(b)	7 per cent sodium chloride	0.7 per cent of total
(c)	9 per cent sodium chloride	0.0 per cent of total
(d)	20 per cent sodium chloride	16.1 per cent of total
(e)	26 per cent sodium chloride	95 per cent of total

give hardness values of 300 to 390 DPH (30 to 40 Rc), while the properly hardened areas will show 685 DPH (60 Rc), readings being taken directly on the etched surfaces. Another distinction often observed between hard and soft areas is that the former are more deeply attacked by the hot etchant, leaving the soft spots raised in relief above the general level of the cylindrical surface.

To determine the area of the steel surface that has failed to harden, the soft spots are first delineated in pencil to render their outlines clearly visible. Then a piece of transparent celluloid on which is ruled a $\frac{1}{8}$ -inch rectangular grid is placed over the curved surface of the specimen. Counting the number of squares over soft areas, the area of soft surface is evaluated in percentage of the whole curved area. The flat end surfaces of the cylinders are excluded from these measurements due to the unknown conditions induced there by the gripping tongs during quenching.

EXPERIMENTAL RESULTS

Some initial experiments were conducted for the purpose of checking various explanations of the cause of soft spots. These ideas are all concerned with control of vapor films. In the first stage of liquid quenching all heat transfer is across a continuous vapor

film and therefore relatively slow. If this film exists for more than a few seconds, hardening is certain to be incomplete; so one thought is that the suppressing of this film will eliminate soft spots. However, even in the second stage of surface cooling on quenching, where the cooling rate is a maximum, soft spots may form due to vapor blanketing of localized areas. There are two distinct types of

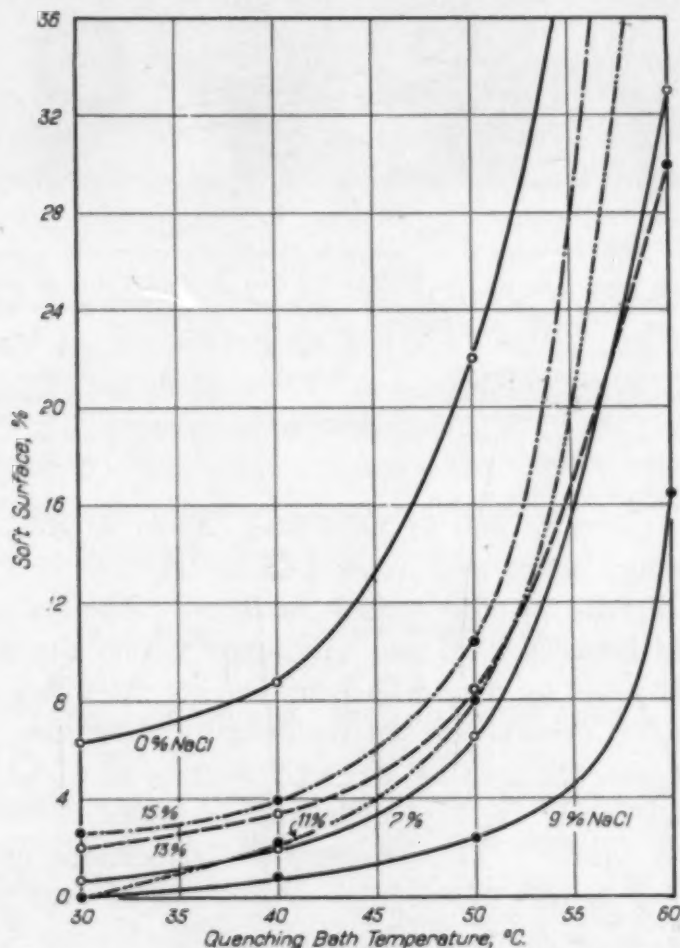


Fig. 3—Relation of Area of Soft Spots on Cylinders of Fine-Grained S.A.E. 1045 Steel to Temperature of Quenching Brines.

soft spots described in the next section, which apparently may be correlated with vapor films in these respective stages of quenching.

Epstein and Rawdon³ attribute soft spots to dissolved atmospheric gases, and Palmer⁴ maintains that the benefit conferred by additions of common salt are due to reduction in solubility of these gases by salt. To test this possibility specimens were quenched in

³S. Epstein and H. S. Rawdon, "Steel for Case Hardening—Normal and Abnormal Steel," Bureau of Standards Journal of Research, Vol. 1, 1928, p. 423.

⁴F. R. Palmer, "Tool Steel Simplified," The Carpenter Steel Co., 1937, p. 141.

tap water that had been degasified by reduction of air pressure over it to a value at which it boiled at room temperature. After so boiling for 6 hours, specimens quenched therein showed 2.0 per cent soft area compared with 2.3 per cent for normal tap water, both tests being made at a bath temperature of 16 degrees Cent. At this temperature there are no soft spots on quenching in a 9 per cent sodium

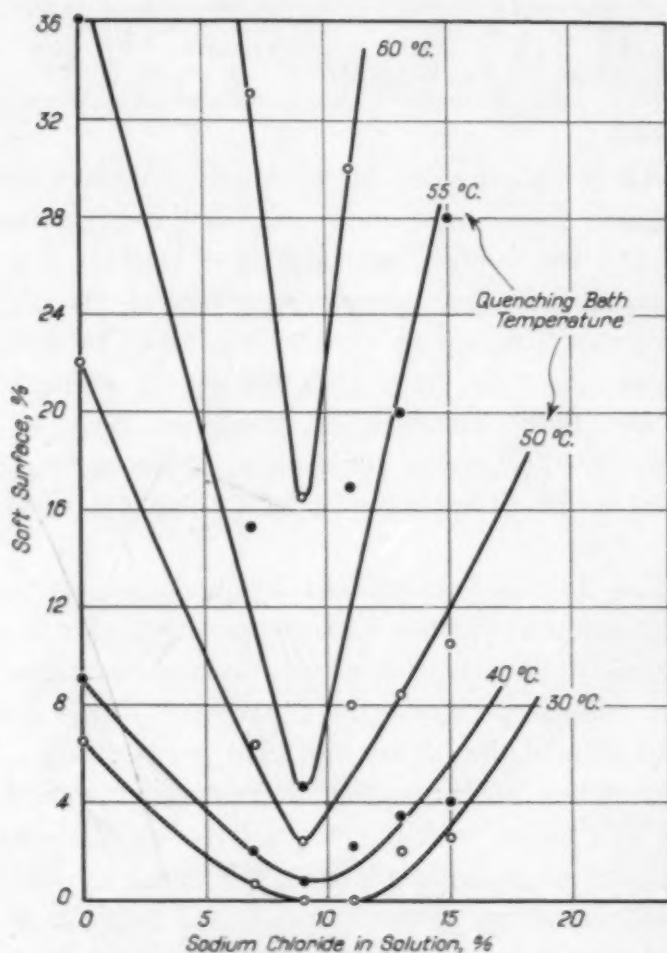


Fig. 4—Relation of Area of Soft Spots on Cylinders of Fine-Grained S.A.E. 1045 Steel to Concentration of Sodium Chloride in Quenching Brine.

chloride brine. This salt content reduces but does not eliminate dissolved gas. Evidently content of normal amounts of atmospheric gas in the water is of little concern and the benefit of salt additions is due to some other factor than elimination of atmospheric gases.

Many other hypotheses advanced for the explanation of the actions of salt additions were investigated in the hope of finding one consistent with known facts, and which would simplify the problem of finding the best addition agent. Failing to find such an hypothesis,

Table II
Area of Soft Spots in Cylinders of Fine Grained S.A.E. 1045 Steel Quenched from
816° C. (1500° F.) in Unagitated Sodium Chloride Brines.

Brine Temperature		Soft area in per cent of total						26% NaCl
°C.	°F.	0	7	9	11	13	15	
0	32	0.6
20	68	3.0
30	86	6.4	0.7	0.0	0.0	1.9	2.6	95
40	104	8.8	2.0	0.7	2.1	3.5	3.8	96
50	122	22	6.5	2.4	8.0	8.3	10.4	..
60	140	55	33	16.5	30	50	41	..

the experimental study of addition agents was undertaken, first attention being given to the most commonly used salt, sodium chloride.

Fig. 2 shows the surface appearance of etched test specimens quenched in brines of various concentrations at 30 degrees Cent. They revealed from 0 to 95 per cent soft area. The individual test results are quite consistent both with regard to temperature variation, Fig. 3, and bath concentration variation, Fig. 4. Table II summarizes these results, giving some data not included in the curves. The detrimental effect of increase in bath temperature is very pronounced.

A surprising fact is brought out by plotting soft area against solution concentration at various bath temperatures (bath isotherms) as in Fig. 4, namely, that there is a well defined optimum concentration of sodium chloride at 9 per cent by weight. This concentration is more sharply defined the higher the bath temperature. Obviously where brine quenches are used the concentration should be maintained close to this value, which is easily done by checking the solution with a hydrometer or salinometer. The density value of a 9 per cent sodium chloride solution in water is 1.063 at room temperature.

Why there should be a well-defined minimum in soft spots with brine concentration is indicated by considering two opposing effects. The one responsible for loss in efficiency must be the reduction in cooling rate within the second stage with increase of concentration, substantial evidence of which had already been quoted in another paper.² Apparently the influence responsible for gain in quenching efficiency is reduction in duration of the first or vapor stage of cooling, also discussed in the same paper. Evidently the effect of the first factor becomes dominant after the optimum concentration is passed, that of the second factor being dominant at lower concentrations.

Another addition agent, caustic soda (NaOH), also used commercially, was tested in a similar manner. An optimum concentration was found for it also, at about 3 per cent and possibly less, Fig. 5. At this concentration it is as effective as sodium chloride. This agent has not been used widely because of its harmful effect on the skin at the higher contents commonly considered essential for effective

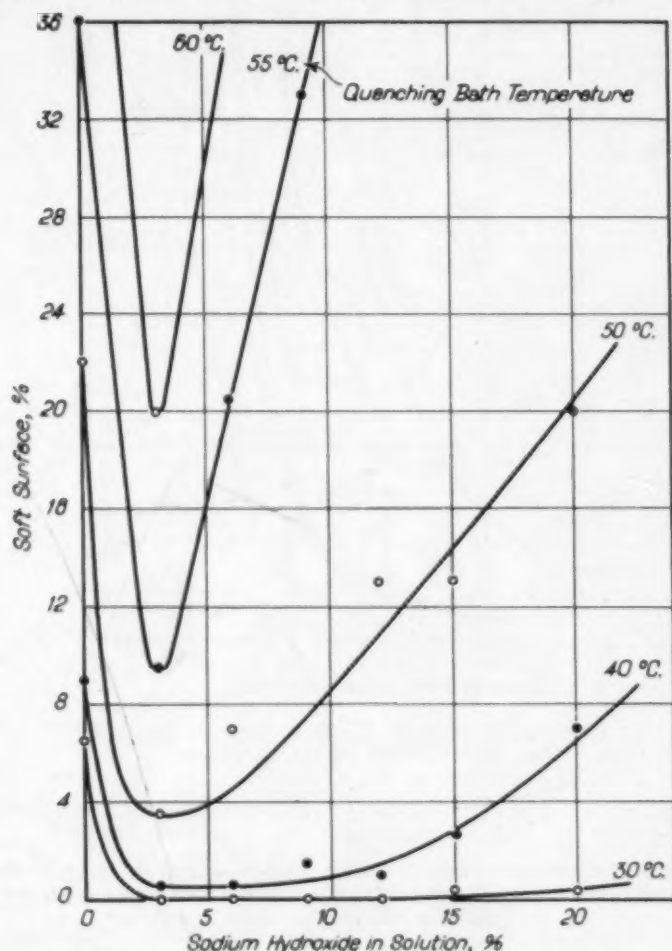


Fig. 5—Relation of Caustic Soda Concentration in Water Solutions to Area of Soft Spots on Specimens Quenched Therein.

action. At 3 per cent, however, no bad effects were experienced from an occasional contact in the laboratory.

The quenching efficiency of a number of other water solutions was determined but no attempt was made to find the optimum concentration of each. Obviously this must be done to determine the real merit of any addition agent. The results are given in Fig. 6. A striking fact brought out by these experiments is that some inorganic solutes have a detrimental effect, for example, borax and

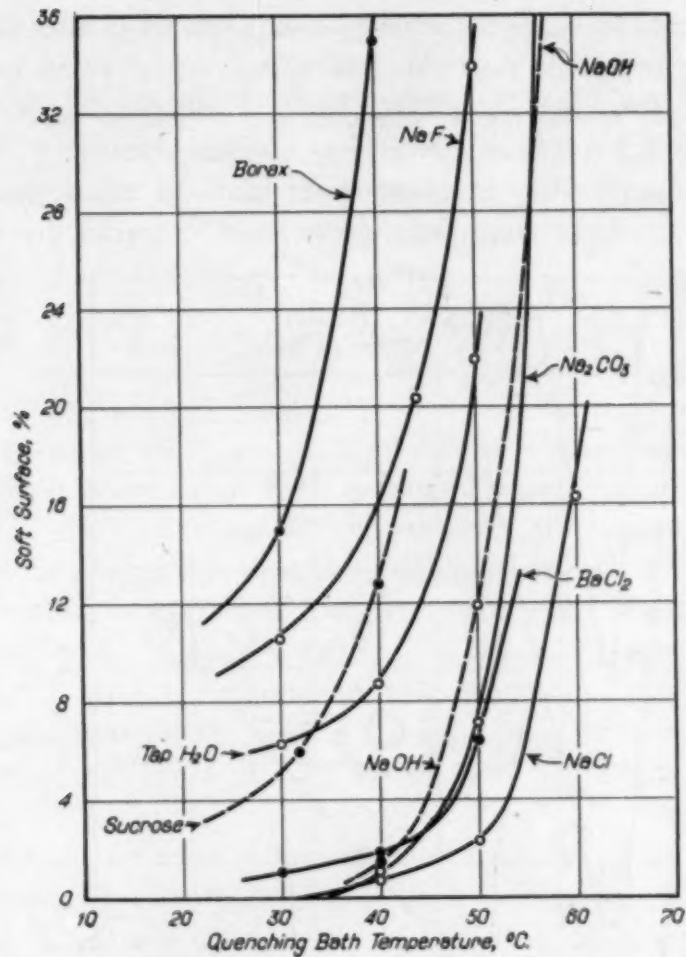


Fig. 6—Relation of Soft Surface Area to Quenching Bath Temperature for Various Water Solutions.

Table III

Temperature of Aqueous Solutions which Produce Soft Spots Over 2 or 10 Per Cent of the Area of Fine Grained S.A.E. 1045 Steel Cylinders Quenched Therein from 816° C. (1500° F.).

Solute	Concentration in Weight %	Quenching bath temp. for:	
		2% Soft Area	10% Soft Area
(Fresh tap water)	0	12	42
(Degassed tap water)	0	16	..
Sodium chloride	7	41	53
Sodium chloride	9	49	57
Sodium chloride	11	40	52
Sodium chloride	13	30	52
Sodium chloride	15	..	50
Sodium hydroxide	3	47	56
Sodium hydroxide	6	45	52
Sodium hydroxide	9	42	49
Sodium hydroxide	12	43	49
Sodium hydroxide	15	38	48
Sodium hydroxide	20	33	44
Barium chloride	9	41	52
Sodium carbonate	9	43	51
Sucrose	9	14	37
Sodium fluoride	4	..	27
Borax	4	..	16

sodium fluoride. From other work it is known that soap is particularly bad because of its effect in greatly extending the first stage of cooling.⁵

These various aqueous solutions can be compared very effectively by tabulating the temperatures at which a given percentage of soft area is obtained. This is done in Table III for 2 per cent and 10 per cent soft area, the values being taken from the curves. The data for common salt and caustic soda are also plotted in Fig. 7 to show how

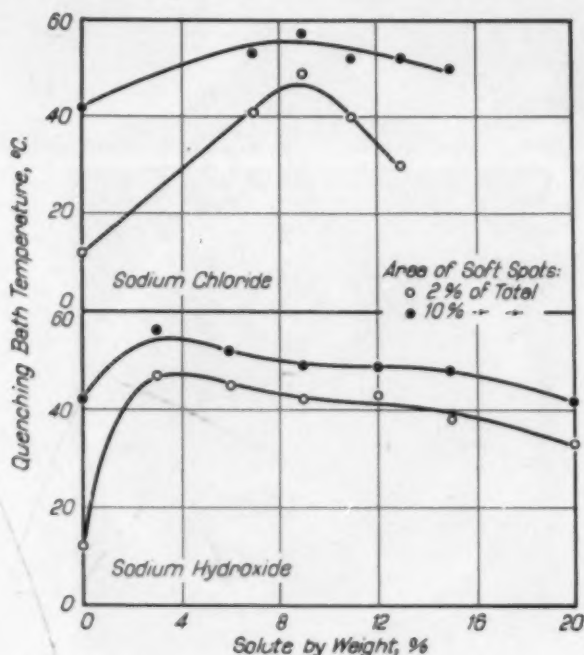


Fig. 7—Effect of Sodium Chloride and Hydroxide Concentrations in Water Quenching Bath on Bath Temperature Required to Produce 2 Per Cent or 10 Per Cent Soft Area.

this characteristic property of an aqueous solution varies with bath composition. The higher the temperature recorded the less danger there is of soft spots occurring on work quenched in the bath at a normal temperature. The increase in this characteristic temperature with a given addition over that for tap water may thus be interpreted as an added factor of safety. The additional latitude relative to that of tap water available with the two best solutions permits hardening without stirring where it might otherwise be required to eliminate soft spots. Again, it permits hardening of larger sections without incurring soft spots.

⁵N. B. Pilling and T. D. Lynch, "Cooling Properties of Technical Quenching Liquids," *Transactions, American Institute of Mining and Metallurgical Engineers*, Vol. 62, 1920, p. 665.

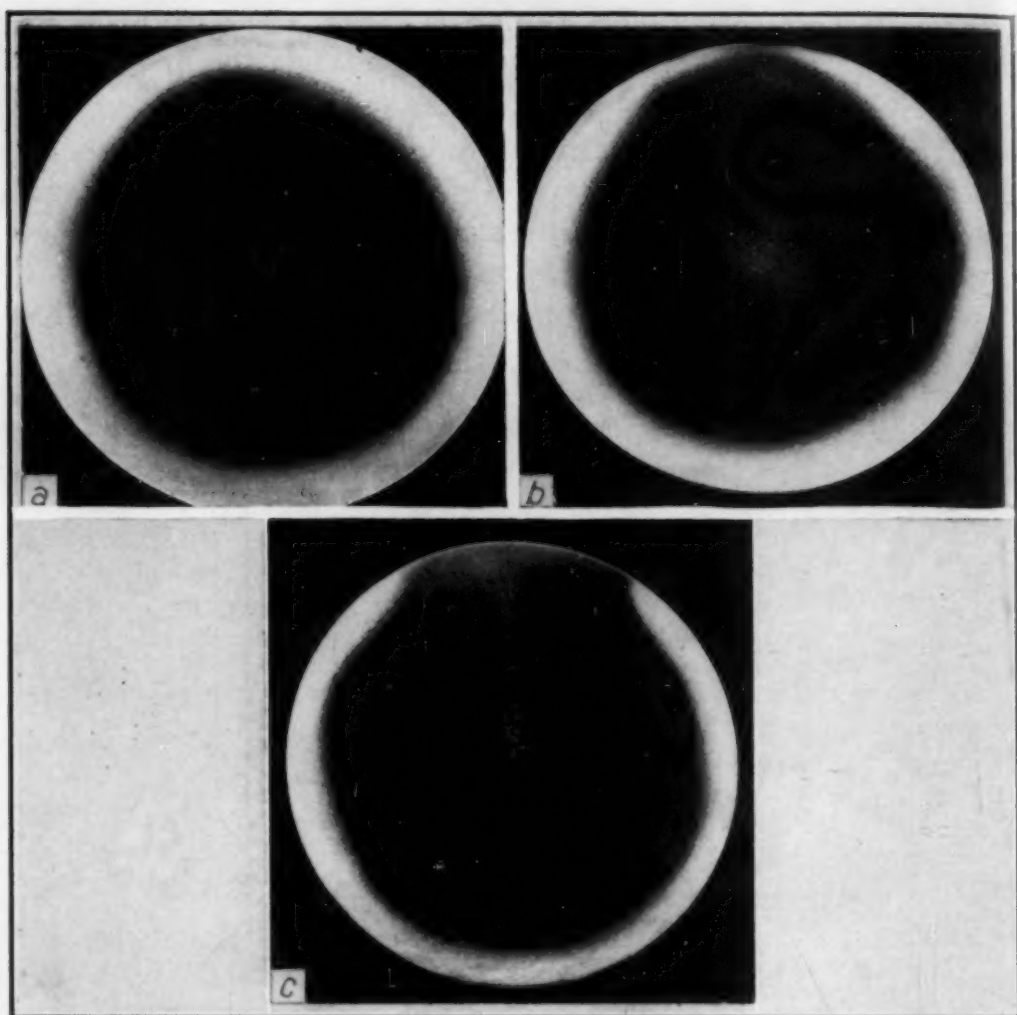


Fig. 8—Etched Sections of Steel Specimens Incompletely Hardened by Quenching From 815 Degrees Cent. (1500 Degrees Fahr.) in Unagitated Tap Water at (a) 0 Degrees Cent. (32 Degrees Fahr.), (b) 20 Degrees Cent. (68 Degrees Fahr.), and (c) 50 Degrees Cent. (122 Degrees Fahr.). Magnification: $\times 2$. (Reduced in Reproduction)

CHARACTERISTICS OF SOFT SPOTS

While the total area of soft spots can be satisfactorily evaluated by surface etching, this preparation tells nothing about their characteristics in a third dimension. To study depth characteristics a number of sections were made through soft spots and prepared by lapping after the technique of Ellinger and Acken.⁶ Etching with 1 per cent Nital darkened the pearlite, leaving the martensite white. Photographs of representative specimens are shown in Figs. 8 and 9.

The most common type of soft spot is that illustrated by Fig. 8c where there is a sharp discontinuity in the martensite shell, the

⁶G. A. Ellinger and J. S. Acken, "A Method for the Preparation of Metallographic Specimens," *TRANSACTIONS, American Society for Metals*, Vol. 27, 1939, p. 382.

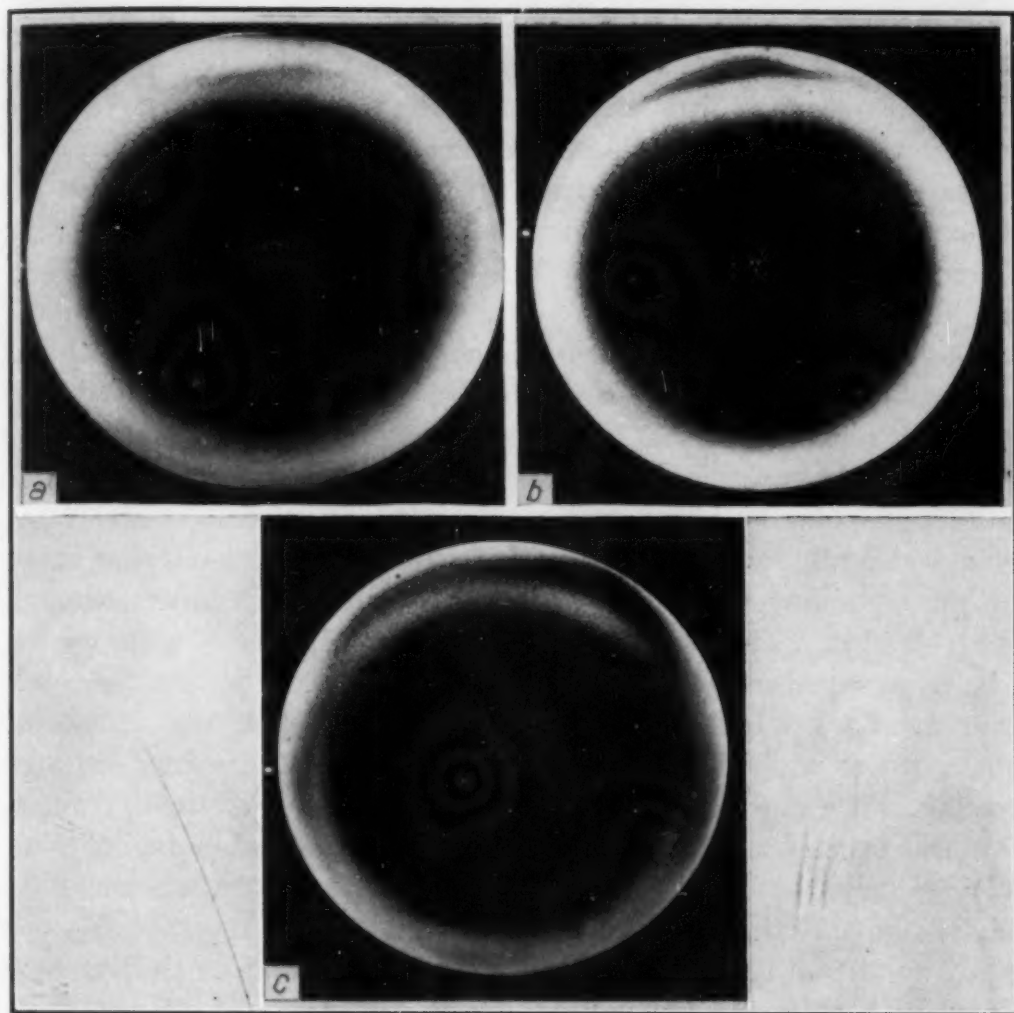


Fig. 9—Etched Sections Through Steel Specimens Incompletely Hardened After Quenching From 815 Degrees Cent. (1500 Degrees Fahr.) in Unagitated Sodium Chloride Brines at (a) 9 Per Cent Concentration, 30 Degrees Cent. (86 Degrees Fahr.) Brine Temperature; (b) 15 Per Cent Concentration, 30 Degrees Cent. (86 Degrees Fahr.) Brine Temperature; (c) 20 Per Cent Concentration, 40 Degrees Cent. (104 Degrees Fahr.) Brine Temperature. Magnification: $\times 2$. (Reduced in Reproduction)

pearlite in the soft core extending continuously to the surface. An abrupt change in hardness occurs in passing over the boundary and the cylindrical surface boundary is well defined by etching. Normal hardening of the type desired is illustrated by Fig. 8a and an intermediate case where penetration of hardening is less shallow, but no pearlite appears at the surface, is shown in Fig. 8b. The latter case would not be identified as a soft spot by surface etching.

Another kind of soft spot was occasionally encountered in which the boundaries are rather vaguely defined by the surface etching and in which the minimum hardness was only slightly below full martensite hardness. In this case the hardness transition is gradual.

These spots nevertheless were recognized by surface etching and included in the soft spot area determinations.

Cross sections through this second type of soft spot revealed why it was much less well-defined than the first type, Fig. 9. Here three manifestations of the same phenomenon are shown. Pearlite islands occur in the martensite shell. The island may be near an inner boundary of the shell as in Fig. 9a, intermediate, (b), or extending to the surface, (c). How near this island comes to the cylindrical surface determines whether or not it is revealed by the surface etch. Thus (a) cannot be revealed by a surface etch or hardness test, (b) can be but is poorly defined, and (c) is as easily recognized as is the first type of soft spot.

Soft spots of the first type occur mostly in the less concentrated brines and solutions containing solutes of detrimental effect. Those of the second type are found almost exclusively in brines of high concentration. This finding suggests at once that soft spots of the first type correlate with the factors responsible for optimum salt concentrations. It has already been indicated that the beneficent effect of salt additions is due to reduction in duration of vapor stage cooling. The first and simpler type of soft spot is evidently caused by local excessive duration of that stage of cooling, a higher bath temperature being required to produce it the higher the concentration of sodium chloride up to 9 per cent.

The second type of soft spot, on the other hand, is clearly associated with reduced surface cooling rate in the second or fast stage of quenching with increase in brine concentration. Further evidence of this effect was obtained by measuring the depth of hardening for 50 per cent martensite in normal specimens at several brine concentrations, Table IV. Penetration of hardening in a given steel under controlled conditions is, of course, a measure of effective surface cooling rate when vapor film cooling is suppressed. The values recorded show that this cooling rate is slowly but definitely lowered by increase in salt concentration.

The slower surface cooling (shallow hardening) on quenching in brines of high salt concentration is actually more favorable to soft spot formation than the faster cooling of water, though soft spots are less frequent. That soft spots are less frequent is due to practically complete suppression of vapor film cooling by salt addition. Another blanketing agent must then be found to account for the occurrence of soft spots with high salt concentrations. That agent is

Table IV
Penetration of Hardening In $1\frac{3}{8}$ Inch Diameter Cylinders of Fine Grained S.A.E. 1045 Steel Quenched in Brines at 20° C. from 816° C. (1500° F.).

Concentration of Sodium Chloride	Depth to 50 Per Cent Martensite
0%	0.145 inch
5	0.145
9	0.142
15	0.138
20	0.135

undoubtedly the vapor bubbles which are responsible for the extremely rapid heat transfer in the fast stage of quenching. These bubbles form very rapidly at the metal surface and must be equally rapidly removed from the surface to maintain the fast cooling necessary for normal hardening. Being moved by gravity they naturally concentrate on the upper surface of specimens quenched horizontally. Nearly all soft spots in specimens of the second kind are found in that position.

By reference to the Bain curve of reaction rates it can be readily seen how vapor blanketing of less than 1 second duration can produce soft spots of the second type as shown in Fig. 9. At a temperature of about 1050 degrees Fahr. (566 degrees Cent.) the rate of reaction is a maximum and pearlite will form in less than 1 second. If now the surface of the specimen has cooled at a normal rate to this temperature and the cooling is interrupted momentarily by a vapor bubble, pearlite will form at the surface as in Fig. 9c. Immediate recovery of surface cooling prevents progress inward of the transformation because at the instant of interruption the metal below the surface is at a higher temperature and hence at one where the reaction rate is much slower. In the same way a small soft spot can form at deeper positions within the martensite shell such as those illustrated by Fig. 9a and (b) if the interruption of cooling occurs when the surface is at lower temperature than 1050 degrees Fahr.

A significant observation in connection with soft spots of the second type is that penetration of hardening is actually deeper under the pearlite island in the martensite shell than under the continuous martensite shell in the same specimen, as may be seen in Fig. 9. Though this seems anomalous, it is consistent with the views expressed here. When the pearlite island forms, it is surrounded by austenite, transformation of the core not starting until the core rim, being deeper, has cooled to 1050 degrees Fahr. or some temperature

close thereto. The pearlite island has much higher thermal diffusivity than the surrounding austenite and therefore provides faster cooling at points under it than would occur in its absence. Faster cooling, of course, produces deeper hardening. Viewed from another angle this is evidence that transformation of pearlite islands in a martensite shell occurs earlier in the quench than first transformation of core pearlite.

SUMMARY AND CONCLUSIONS

Though water solutions are the most active of all known quenching media, when unagitated as is necessary to minimize distortion, they fail to eliminate consistently the formation of soft spots in shallow hardening steels. As a means for studying and evaluating their quenching efficiency in this regard, a test method was developed based on the quantitative measurement of the area of soft spots, formation of which is deliberately induced by heating the quenching bath. Merit of a solution tested is shown to be measured by the bath temperature at which soft spots first appear or are present to a given small degree.

Applying this test the following information on aqueous quenching baths was obtained.

(1) Dissolved atmospheric gases in Pittsburgh tap water have an inconsequential effect on the occurrence of soft spots.

(2) Moderate additions of common salt to water reduce materially the incidence of soft spots and reduce slightly the penetration of hardening.

(3) There is a well-defined optimum content of sodium chloride at 9 per cent by weight.

(4) Caustic soda acts in the same manner as common salt except that the optimum concentration is 3 per cent by weight or possibly somewhat less and the concentration less critical.

(5) Contents of sodium chloride and hydroxide near saturation are decidedly harmful as is 4 per cent of borax or sodium fluoride.

(6) Two types of soft spots are distinguished, namely, one in which the surface pearlite is continuous with core pearlite and the other in which an island of pearlite forms in the martensite shell.

(7) Penetration of hardening is materially deeper under a pearlite island than where hardening is normal.

(8) Soft spots continuous with core pearlite are due chiefly to extended duration of the first or vapor stage of liquid cooling.

(9) Momentary vapor blanketing in the second or fast stage of liquid cooling is the apparent cause of pearlite islands in martensite, the blanketing occurring when the island area is at the temperature of the peak of the Bain reaction rate curve.

(10) That there is an optimum concentration of sodium chloride and hydroxide is attributed to the fact that the beneficial effect of increase in solute concentration, i.e., shortening of the vapor stage, is opposed by the detrimental effect of reduced cooling rate in the fast stage.

DISCUSSION

Written Discussion: By Gordon T. Williams, Deere & Co., Moline, Illinois.

A very useful test procedure is here presented, as is to be expected from the fine previous contributions by Mr. Scott. The writer was especially interested in the results obtained on sodium carbonate solutions for two reasons. First, Na_2CO_3 is gradually formed in a sodium hydroxide solution used in an open tank; the writer was under the delusion that Na_2CO_3 was damaging to the quench and it is well to be disabused of this notion. From the data given for Na_2CO_3 (omitted, incidentally, from the curves of Fig. 6) its effect must be comparable with NaOH and hence CO_2 absorption is not damaging. Secondly, perhaps sodium carbonate would be found on thorough investigation such as given NaCl and NaOH to be just as effective in proper concentration, and a very useful medium made available which would be free from tendency to promote rust (such as occurs with NaCl) and the personal hazards of NaOH.

Can the authors discuss effect of contaminants resulting from use such as finely suspended iron oxide, on the quench; that is, give some basis for discarding used solutions whose salt content is still satisfactory?

Written Discussion: By Charles M. Lichy, Jones & Laughlin Steel Corp., Pittsburgh.

The authors have uncovered some interesting data in their development of a test method for the evaluation of aqueous quenching media for steel hardening. However, after studying their testing procedure, there are two questions which arise and appear to need some clarification.

We are wondering if the authors have not minimized too greatly the effect of scale formation on their test pieces and the explosive elimination of same in the quenching media. While they mention that no scale remained on the fully quenched samples, we have found that different types of scale formed during a heating operation require varying degrees of time to be removed. This is true even though all scale is completely eliminated during the quenching period. We believe that it is this hindrance of quenching during the first or vapor stage of liquid cooling that causes soft spots in those areas where the scale was tightest. These observations were made with a heat

of fine-grained S.A.E. 1045 steel which was very susceptible to soft spots under ordinary heating and quenching conditions. Samples of bar stock from this heat were washed in a warm (approximately 180 degrees Fahr.) 50 per cent solution of hydrochloric acid before being placed into a furnace. The result was that a fine flaky scale was produced which was removed completely and instantaneously upon quenching. All bars treated in this manner failed to show any indications of soft spots. May we therefore ask if it is not possible that because of this scale condition the optimum solution of both sodium chloride and hydroxide is that solution that will allow by explosive action the fastest removal of scale in the still quench?

Further, with a still quench as used in this experiment, there is no propulsion of the liquid by external force past the hot bar or sample that was quenched. The only motion of the liquid was due to natural convection, or the mechanical convection set up by the difference in density of the liquid at various points. This rate of circulation varies widely with the viscosity of the liquid and temperature difference. As the warming or heating of the bath proceeds, both temperature difference and viscosity of the liquid decrease. Both these variables have an opposite effect. It is, therefore, possible that a solution may be a very effective quenching agent but because of density or viscosity does not allow a rapid enough convection of heat to allow any great depth of hardness penetration. The data on page 529, Table IV, would tend to confirm this statement, in that a minimum of soft spots are produced but yet the depth of hardness is also decreased. From this is it not possible that some form of agitation or swift movement of a quenching media would tend to change the optimum concentrations of the quenching media?

We are, therefore, offering for further discussion and consideration the effect of scale conditions and the possibility of a change in optimum solutions if the quenching bath were agitated or the media effectively circulated.

Written Discussion: By C. L. Harvey, The Lamson & Sessions Co., Cleveland.

The authors of this paper have made an interesting and valuable contribution to our knowledge of aqueous quenching media. Their technique in treating the specimens follows closely commercial heat treating practice and the quantitative results obtained are therefore of considerable practical value. Figs. 4 and 5 showing the results obtained with various concentrations of sodium chloride and sodium hydroxide are of especial interest. The optimum point of 9 per cent for the chloride quench is so pronounced that I feel we will all watch our salt concentrations more closely when quenching into a still bath.

There are several questions that I would like to ask:

First—Have the authors tried steels of other carbon content to determine whether the optimum solute point is shifted?

Second—With a mildly agitated bath, would not a concentration lower than the optimum produce freedom from soft spots and at the same time greater depth of hardness?

Written Discussion: By B. R. Queneau, instructor of metallography and ferrous metallurgy, Columbia University, New York.

The results obtained by the authors in evaluating different quenching media

are interesting. At first it would seem unsatisfactory to test aqueous solutions at temperatures much above 25 degrees Cent. when attempting to evaluate their properties at room temperature, but from the consistent curves obtained for different media over a wide range of temperatures, it appears that this is not a valid objection.

Since the effectiveness of the quench has been shown to be greatly affected by the composition of the quenching solution, it is well to bear in mind that it is necessary to test a solution over a wide range of concentrations before it is possible to definitely reject an addition agent as unsatisfactory.

In any water quenching operation, one important factor is that of the surface of the steel to be quenched. The authors state that no indication of interference by scale with quenching action of the liquids was observed and



Etched Section Through
Steel Specimen. $\times 1$.

therefore they decided to assume the condition of the surface of the steel to be a constant. Yet we know that the removal of scale occurs much more rapidly in brine solutions than in water. The excellent photographs, Figs. 1 and 2, taken by Prof. I. N. Zavarine of Massachusetts Institute of Technology (*Metal Progress*, April 1935) of steel samples quenched into water and brine demonstrate this point. When the scale does start to come off in a water quench, it does not come off uniformly over the whole surface but tends to stay on in spots and this can undoubtedly be the cause of soft spots. It would have been interesting to have used both scaled and scale-free samples in the testing of aqueous solutions to see how much influence the presence of scale had on the relative efficiency of the different quenching media.

The authors' photographs of etched sections through incompletely hardened steel specimens give an excellent picture of the various shapes of soft spots. However, is it necessary to divide these soft spots into two distinct types? Contrary to the findings of the authors who obtained soft spots of their second type only in solutions of high salt concentration, I have frequently noticed such soft spots in samples that had been quenched in water. An unusual soft spot of this type is shown in Fig. 3. The steel sample, a $1\frac{1}{2}$ -inch round, had been partly packed in cast iron chips and a spot of light

scale formed on the exposed surface. The soft spot was found immediately below this scale.

It will be noted that martensite is present to a depth of 0.135 inch below the soft spot while the martensitic case is only 0.100 inch in thickness. To have martensite present below the soft spot at a greater depth than found elsewhere, the cooling rate past the 1000 degrees Fahr. temperature range must have been greater than at other points of similar depth in the sample. The explanation of the authors that this is due to the greater thermal diffusivity of the pearlite appears to be satisfactory.

To explain the occurrence of the two pearlite bands shown in Fig. 3, it is necessary to consider the scale spot which we know was present on the surface. The scale decreased the cooling rate of the austenite sufficiently to cause it to transform to pearlite even at the surface. The quenching action then removed the scale and the resultant rapid cooling rate caused the formation of martensite. The inner pearlite band was formed by an interruption in the cooling by the momentary formation of a vapor film at the oxidized surface.

The presence of scale on a steel sample may not be directly responsible for the formation of soft spots, but any condition, such as an oxidized surface, which will cause differences in the ability of the surface to maintain a vapor film will affect their formation.

Authors' Closure

Mr. Williams' interesting conjecture regarding the relative merits of NaOH and Na_2CO_3 as addition agents was based on the limited data presented on Na_2CO_3 in Fig. 6 and Table III, and he correctly pointed out that more complete tests would be necessary to determine the value of Na_2CO_3 solutions. Such tests have since been made under the test conditions outlined in the paper at a bath temperature of 30 degrees Cent. with the results tabulated below:

Per Cent Solute	Per Cent Soft Spots
0	6.4
2.2	2.9
6	0.7
10	5.4
14	16.7

Whereas the addition of sodium carbonate up to about 6 per cent by weight evidently improves the quenching efficiency of water, the gain is not so great as is secured with optimum additions of NaCl and NaOH. However, starting with optimum NaOH concentration the absorption of CO_2 will not necessarily improve quenching efficiency.

No tests have been made to determine the possible effect of suspended iron oxide on the prevalence of soft spots.

In regard to Mr. Harvey's first question, no tests were made with steels other than the fine-grained S.A.E. 1045 composition used because we believe that the phenomena investigated were properties of the quench only, and thus

independent of the particular test steel composition, provided only that the test steel is sufficiently susceptible to soft spot incidence to give measurable effects when quenched into the better quenching media. The answer to his second question is yes, but the advantage gained in hardness penetration by substituting agitation for salt content is at most that from 0.142 to 0.160 inch penetration to 50 per cent martensite.

Both Messrs. Lichy and Harvey raise the question of the effect of circulation or agitation of the bath on the optimum solute content. It is known that sufficiently active agitation, as, for instance, that obtained in spray quenches, will completely suppress the formation of soft spots. In other words, there is no obvious need for any addition to water with very active agitation. Our objective in these experiments was to learn how to quench water hardening steels effectively without agitation so as to minimize distortion.

Mr. Lichy's concept of cooling by natural convection appears to describe conditions during the "C" stage better than those during the "B" stage of quenching. The extremely high cooling power of water in the "B" stage is unquestionably due to its high heat of vaporization. The heat stored in vapor formed at the metal surface is rapidly moved from that surface by the large density differential between vapor and liquid, and dissipated by condensation at a distance. This process is, of course, an extreme manifestation of natural convection, and would not be so effective were the viscosity of water very high.

An alternate hypothesis concerning the action of salt in water is well presented by Mr. Lichy, who maintains that soft spots are caused by adherence of scale which does not flake off until late in the quench; Mr. Queneau supports this view by citing evidence that scale is thrown off more quickly in brine than in water. Thus, both Mr. Lichy and Mr. Queneau observe a correlation between soft spots and time at which scale is thrown off. Our observations did not cover this point, though one should expect such a correlation from the vapor film hypothesis of soft spot formation.

That soft spots occur when flaking is delayed arises from the fact that the quenching liquid must wet the scale before it can penetrate; penetration, of course, being necessary for the explosive elimination of scale, identified here as flaking. Wetting terminates "A" stage, cooling so the time at which scale elimination occurs is a measure of the duration of "A" stage cooling, and, therefore, of the extent of soft area. The observations cited are, therefore, in full accord with our view, though their interpretation is not. If scale type and location were the dominant factor, one could not obtain a consistent correlation between soft spot area and bath temperature, or other bath variables. Also, soft spots would not be found more prevalent on the top side of horizontal specimens than on the bottom. We maintain, therefore, that vapor film, except in extreme cases, is the dominant factor in soft spot incidence.

A very heavy scale can, of course, inhibit hardening of the metal directly under it during water quenching. There is, however, evidence also that a thin film is beneficial. Thus, Sato* reports uniform hardening of a carbon steel on

*Science Reports, Tohoku Imperial University, Vol. 21, 1932, page 564.

water quenching only when the work was covered with a thin film of ceramic called "Tonoko." Though our repetition of this experiment failed to confirm it, there is abundant evidence that a highly polished and bright surface is quite unsatisfactory. The persistence of the spheroidal state when a drop of water falls on a heated platinum dish is a good example. Oxidized steel surfaces (film thickness about 0.0002 inch) are wet by a drop of water much more readily than a bright surface. It is, therefore, quite certain that the duration of "A" stage cooling, that is, the spheroidal state, is much less in the presence of an oxide film than in its absence. The bad effects of a heavy scale must, therefore, be limited to the "B" stage of cooling. In this respect the effect of oxide films on steel hardening under water quenching is analogous to that of salt additions to the quenching bath. We do not suggest, however, that control of surface condition alone is a practical means for eliminating soft spots in carbon steels. In fact, there is probably little to be gained in uniformity of hardening by surface preparation over the results obtained in our experiments.

Several of the points brought up by Mr. Queneau have already been considered in preceding paragraphs. With regard to his question as to the necessity for distinguishing two types of soft spots, there is none other than rhetorical convenience. Soft spots of the second type so designated are found by him after plain water quenching, though we have noted them only after quenching in highly concentrated brines. No inference is intended, however, to the effect that they are restricted to such brines, though a greater prevalence may be expected because of slower cooling in the "B" stage. A striking example of this type of soft spot is shown by Mr. Queneau to have originated under an oxidized spot, but, as conceded, the oxide could have been responsible for only one of the cooling arrests. To produce this type of soft spot a double cooling arrest is necessary which required the co-operation of vapor to form the second, even if it were not responsible for both arrests.

The authors are gratified by the pertinent questions and comments received which have increased available knowledge of the subject and helped to crystallize the issues involved. Many new and useful experiments are indicated, and we hope to make at least some of them.

KINETICS OF THE AUSTENITE TO MARTENSITE TRANSFORMATION IN STEEL

BY ALDEN B. GRENINGER AND ALEXANDER R. TROIANO

Abstract

The results of a microscopic study of the kinetics of martensite formation in plain carbon steel are herein reported. The method of study provides a reliable direct-evidence survey of the entire process of martensite formation in eutectoid and hypereutectoid steel. Conclusions apply only to eutectoid and hypereutectoid steel, though it is believed that the behavior of steels containing less than about 0.6 per cent carbon differs from that of higher carbon steel only in degree. From the results of this study it is concluded that the "S-curve" of austenite decomposition gives an incorrect picture of the martensite process; hence, these curves are in error for all temperatures within the martensite range of iron-carbon alloys. The older concept that martensite forms from austenite only during cooling through the proper temperature range (advocated, though never proved, by several early investigators) has been found to be substantially correct.

WHEN a specimen of eutectoid steel is quenched with sufficient velocity from the austenite range to room temperature the normal pearlite reaction is completely suppressed and microscopic examination shows that the specimen consists almost entirely of martensite and has little retained austenite, whereas a specimen of 1.75 per cent carbon steel similarly quenched to room temperature will be almost all austenite with little martensite. Large variations in cooling velocity apparently have little or no effect on these resulting structures, so long as the critical cooling velocity is exceeded.

These simple facts have been known for years, and it is natural that repeated attempts should have been made to answer the question of when and how this transformation from austenite to martensite occurs; study has been intensive since 1917 when Dejean

Much of the material presented in this paper has been abstracted from a thesis submitted by A. R. Troiano in partial fulfillment of the requirements for the degree of Doctor of Science, Harvard University, June, 1939.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, Alden B. Greninger is assistant professor of metallurgy, graduate school of engineering, Harvard University, Cambridge, Massachusetts, and Alexander R. Troiano is assistant professor of metallurgy, University of Notre Dame, Notre Dame, Indiana. Manuscript received June 5, 1939.

(1)¹ first demonstrated the phenomenon of "split critical points" in plain carbon steel. None the less, our principal question remains unanswered; there is, at present, disagreement among metallurgists concerning the manner in which the martensite transformation is influenced by the two heat treating variables, temperature and time.

In the following pages an attempt will be made to summarize and contrast the two principal interpretations given to the austenite-martensite transformation, and to list the more important isolated experiments concerned with the kinetics of martensite formation.² A description will be given of the results of our microscopic study of the austenite-martensite transformation in steels of various carbon contents.

THE S-CURVE INTERPRETATION

In 1930 Davenport and Bain (4) first published their "S-curve" representation of the manner in which austenite decomposes at sub-critical temperatures. The original S-curves were plotted (per cent decomposition on a time-temperature graph) from data obtained by the use of the microscope and the dilatometer. Small specimens were quenched from the austenite region into molten metal baths held at various constant temperatures. From temperatures of approximately 700 to 300 degrees Cent. (1290 to 570 degrees Fahr.), the progress of decomposition of austenite was evaluated microscopically on specimens quenched from the metal baths to room temperature after definite intervals of time. The amount of austenite decomposition could be estimated visually because the product of decomposition appeared as a dark-colored constituent in a white matrix. This white matrix represented what had been austenite at the temperature of decomposition, but actually contained a mixture of austenite and white martensite; the white martensite was formed during the final quench to room temperature. For temperatures below about 300 degrees Cent. (570 degrees Fahr.), the progress of transformation was followed by the use of a quenching dilatometer, and the curve of dilation versus time was taken to be a quantitative measure of decomposition versus time for all bath temperatures below about 300 degrees Cent. (570 degrees Fahr.). In 1937 a detailed S-curve for austenite decomposition in a eutectoid steel was

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

²A summary of the literature on martensite has been presented by Epstein (2); most of the important publications that appeared before 1935 will be found listed and discussed in his book. Carpenter and Robertson (3), in their recently (1939) published treatise on physical metallurgy, devote some sixty pages to a discussion of various aspects of the martensite problem.

published (5) as a contribution from the Research Laboratory, United States Steel Corporation. Apparently, the decomposition rates used in plotting the latest curve were evaluated by microscopic methods for all temperatures (few details are given) and this curve differs somewhat from those previously published by Davenport and Bain.

Briefly, the S-curve states that if steel (of eutectoid composition, for example) is "hot-quenched" to any temperature below A_{e1} 100 per cent austenite can be retained for any bath temperature, and a measurable time interval is required for the beginning of austenite decomposition at that temperature. In other words, austenite decomposition is pictured as taking place entirely at constant temperature after the specimen has reached the temperature of the quenching bath; and there are two temperature ranges within which austenite decomposition proceeds with high velocity at constant temperature—one centered around 600 degrees Cent. (1110 degrees Fahr.), where the product is primary troostite, and one for temperatures near or below 50 degrees Cent., where the product is martensite. In eutectoid steel, their measured time interval for the beginning of austenite decomposition reached a maximum (about five minutes) at approximately 250 degrees Cent. (480 degrees Fahr.).

According to the 1937 S-curve, the decomposition of eutectoid steel at constant temperatures (below 300 degrees Cent.) proceeds about as follows: If the steel (small pieces) is quenched into a metal bath held at 200 degrees Cent., the specimen consists of 100 per cent austenite that remains unaltered for a period of 2.5 minutes; decomposition then starts, and ends in 11 hours. Similarly, if the quenching bath is at 100 degrees Cent., the quenched specimen supposedly consists of 100 per cent austenite until a time of five seconds has elapsed; decomposition then begins, and ends in fifteen minutes. If the quenching bath is maintained at temperature levels below about 50 degrees Cent., all austenite is supposedly retained, but transforms in less than two seconds to martensite after the bath temperature has been reached.

CARPENTER AND ROBERTSON'S INTERPRETATION

During the years 1929-1932, Carpenter and Robertson and co-workers (in England) published several papers describing studies of martensite formation in steels of various carbon contents; par-

ticularly important in the group is a paper by Robertson (6) describing a microscopic study of the subcritical decomposition of austenite in steels of eight different compositions ranging from 0.10 to 1.73 per cent carbon. In many ways his studies paralleled those of Davenport and Bain; however, Robertson's object was evidently the assimilation of data that would aid in understanding the process of austenite decomposition at various temperatures, rather than a quantitative evaluation of decomposition rates; and his study of any one composition was not so detailed as that of Davenport and Bain.

Carpenter and Robertson have recently (1939) outlined their interpretation of the austenite-martensite transformation (3). Their ideas can best be summarized, perhaps, by the following quotation taken from their writings:

"In a given steel, the progress of the austenite to martensite change depends entirely on the metal being cooled through the range of temperature between the appropriate points on the lines AM and AM'. [See Fig. 1.] No matter how rapidly the metal is cooled through this range the com-

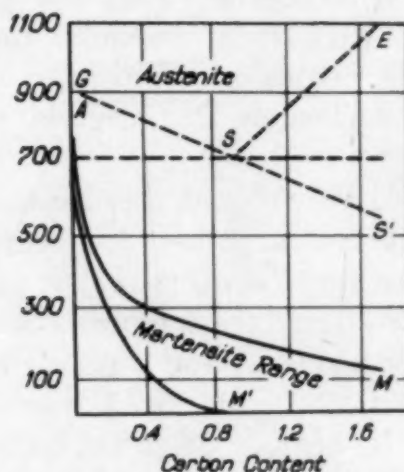


Fig. 1—The Martensite Range in Iron-Carbon Alloys, From Carpenter and Robertson (3). AM Represents the Approximate Temperature at Which the Formation of Martensite Begins on Cooling; AM' Indicates Temperatures at Which the Austenite to Martensite Change is Substantially Complete.

plete transformation from austenite to martensite takes place. If cooling is retarded, the rate of formation of martensite is retarded; if cooling stops, the formation of martensite stops. It is for this reason that in high carbon steels austenite may be retained indefinitely at atmospheric temperature and yet begin to change to martensite whenever it is cooled below

atmospheric temperature. For the same reason, austenite may be retained for a considerable time if cooling is stopped just above the line AM or at any point between the lines AM and AM'.³

"As no rate of cooling can suppress the martensite change, the decomposition of austenite at constant temperatures below this range [below line AM', Fig. 1] cannot be studied, and at temperatures within it only part of the decomposition can be studied at constant temperature."⁴

In commenting upon Davenport and Bain's results for 0.78 per cent carbon steel, Carpenter and Robertson state:

"At lower temperatures [below 180 degrees Cent.] the time required for the complete decomposition of austenite appeared to decrease again, but these were either within or below the martensite range, and the transformation from austenite to martensite took place either partly or completely with great rapidity while the steel was cooled to the constant temperature at which its behavior was to be studied."⁵

MISCELLANEOUS STUDIES

Quenching Curves—Several investigators have measured the effect of carbon content and cooling velocity on the temperature of the Ar" thermal arrest.⁶ The most recent (1937) cooling-curve determination of Ar" was made by Wever and Rose (7), using an improved recording technique. They found that for all carbon contents above 0.4 per cent, cooling velocity had no effect on the temperature of the Ar" thermal arrest. A few of the invariant points located were as follows: 0.6 per cent carbon, 250 degrees Cent.; 0.8 per cent carbon, 200 degrees Cent.; 1.4 per cent carbon, 110 degrees Cent. The results of Wever and Rose are in agreement with most of the earlier studies,⁷ particularly the detailed study of Wever and Engel (8).

Direct Measurement of Martensite Formation—Direct quantitative measurement of martensite formation while a specimen is cooling, and continuation of this measurement after the temperature drop is interrupted, is experimentally difficult to carry out, especially for wide ranges of cooling velocities. Most of the attempted meas-

³Reference 3, p. 882.

⁴Reference 3, p. 892.

⁵Reference 3, p. 893.

⁶See Epstein, Reference 2, for a discussion of early work on this subject.

⁷However, Epstein reports (Ref. 2, p. 172) that unpublished results of time-temperature cooling curves from the National Bureau of Standards show that Ar" in 1 per cent carbon steel is depressed from 270 degrees Cent. to about 140 degrees Cent. as the cooling velocity at 300 degrees Cent. is increased.

urements have yielded incomplete or qualitative results, and are liable to criticism of one form or another.⁸ Experimental techniques are obviously simplified when measurements can be started on austenite specimens at or slightly above room temperature, and several reports have been published describing the manner in which martensite forms when retained austenite is cooled to subatmospheric temperatures. Among these are the microscopic studies of Hane-mann and Wiester (9), the dilatometric studies of Tammann and Scheil (10) and of Scheil (11), and the acoustic studies of Förster and Scheil (12). Förster and Scheil measured acoustically the formation of martensite in specimens of 29 per cent nickel steel cooled to subatmospheric temperatures. They state that martensite formation began at -31 degrees Cent. and continued during cooling; when the cooling was stopped, the transformation did not cease immediately (as was observed in a previous study on plain high carbon steel by Scheil) but rapidly died away. Also, when cooling was resumed, martensite did not begin to form immediately, but did begin at a somewhat lower temperature; in this respect, the observed behaviors of plain high carbon steel and nickel steel were similar.

In general, the above investigators appear to agree⁹ that at least in the low temperature range ($+100$ to -200 degrees Cent.) martensite forms only while the specimen is being cooled and rapidly ceases to form if the temperature drop is interrupted. Obviously, there is a discrepancy between the S-curve interpretation of the austenite-martensite transformation and the more generally accepted idea that martensite forms only during cooling through the proper temperature range. It is true, however, that the direct measurements of martensite formation that have been made (for example, those of Scheil and of Wiester) do not offer any objective disproof of the S-curve interpretation. These measurements have all been made during slow cooling, and it is the contention of the S-curve that the martensite transformation can be depressed to lower temperatures by increasing cooling velocities. Furthermore, there is no reason to assume that the martensite transformation should proceed in the same way at temperatures near 200 or 300 degrees Cent. that it does near atmospheric temperature or below.

⁸Particularly vulnerable (for temperatures below about 200 degrees Cent.) are the recent magnetometric evaluations of reaction-rate curves for 1.16 per cent carbon steel by Wever and Hänsel (13).

⁹Dehlinger's representation (14) of the transformation velocity of iron-nickel alloys at low temperatures as a function of time and temperature has been criticized by Wassermann (15).

It is clear that an attempt should be made to adjust the differences in opinion that have been expressed on this subject. The two prominent interpretations are, in fact, so divergent that it should be possible at least to determine which of the two more nearly approaches reality.

EUTECTOID STEEL

Experimental Methods

In the present study, the method of analysis is based on the fact that tempered martensite and untempered martensite contained in the same specimen are easily differentiated microscopically after nital etching (picral is unsuited for this task) because the former etches dark brown and the latter is almost unattacked and remains white. In principle, this technique is not at all new; it has been used extensively, particularly by Hanemann (9).

The following details refer to the standard procedure used. Departures from this procedure were made only in evaluating the effect of certain variables (such as austenite grain size, tempering treatment, alloy impurities); these departures will be described in later pages.

Material—Steel ingot kindly supplied by Mr. G. V. Luerssen of the Carpenter Steel Company had the following composition:

C	Mn	P	S	Si	Cr	Ni
0.82	0.04	0.004	0.044	0.02	0.07	0.07

Lattice parameters were austenite, $a_0 = 3.581 \text{ \AA}$; martensite, $c/a = 1.038$, both $\pm 0.001 \text{ \AA}$. Measured on small specimens (see below) quenched from 900 degrees Cent. to room temperature in 10 per cent NaOH solution.

The specimen sizes were, (a) Large: $\frac{1}{4} \times \frac{1}{4} \times \frac{5}{32}$ inch and (b) Small: $\frac{1}{4}$ inch long and 0.035 to 0.045 inch square in cross section.

Solution Anneal—Ten minutes at 1200 degrees Cent., cooled to 950 or 900 degrees Cent. (1740 or 1650 degrees Fahr.) and quenched. The furnace consists of a platinum plus 20 per cent rhodium resistance vertical tube with automatic temperature control. Specimens suspended with 20-gage iron wire; small specimens attached to iron wire through about $\frac{1}{4}$ inch of twisted 30-gage chromel wire. The atmosphere was a continuous flow (pressure about $\frac{1}{2}$ inch of mercury) of nitrogen gas, dried, purified over fine copper gauze at about 900 degrees Cent., redried with Anhydrone.

The quenching bath consisted of about five pounds of the quater-

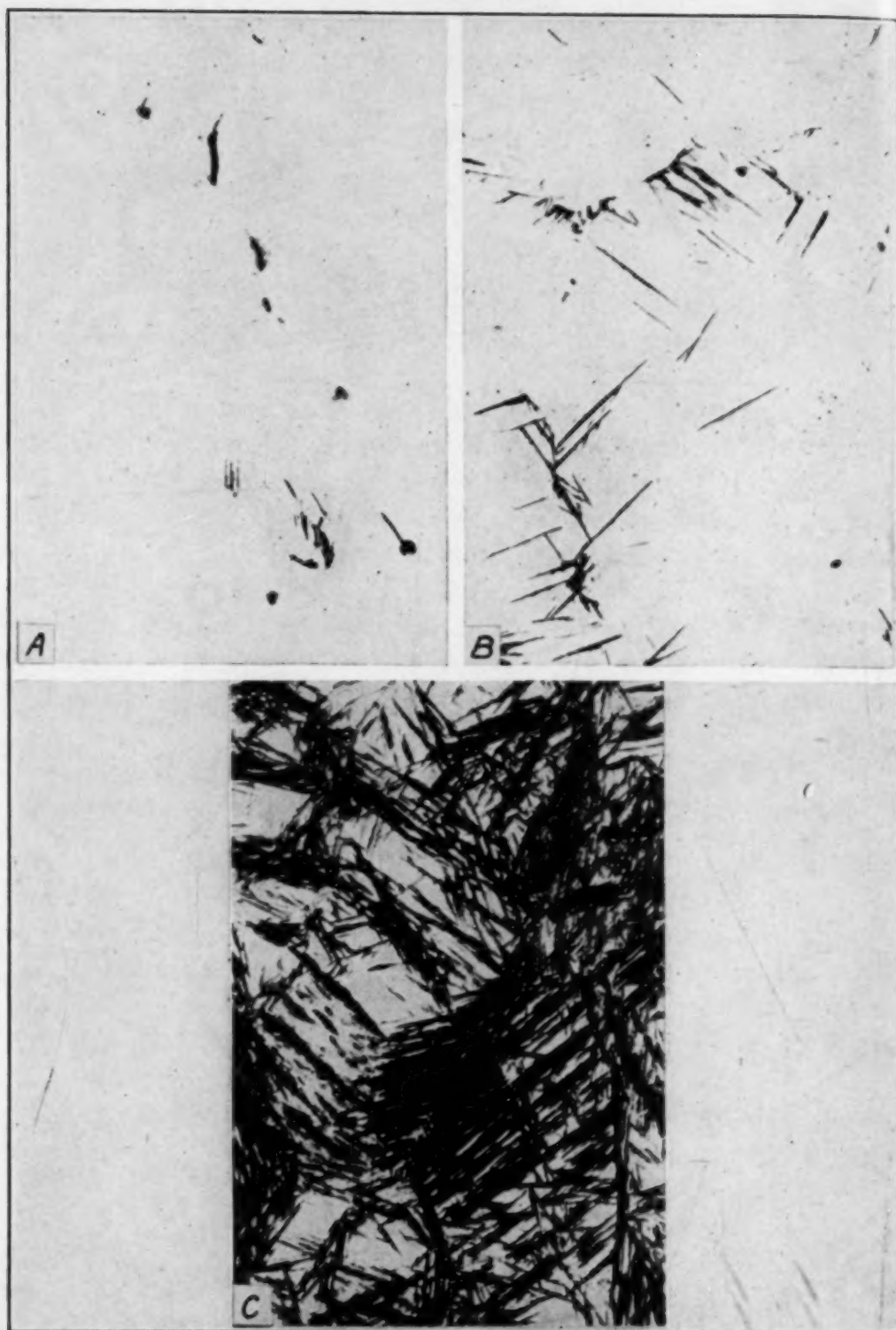


Fig. 2—0.82 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. See Text for Treatment Details. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Nital. White Background Represents Austenite Not Decomposed at Temperature of Run. White Areas Contain a Mixture of Martensite and Austenite; the White Martensite is Formed During Final Quench to Room Temperature.

Fig. 2A—260 Degrees Cent.—2 Minutes. Fig. 2B—260 Degrees Cent.—3 Minutes. Fig. 2C—260 Degrees Cent.—10 Minutes.

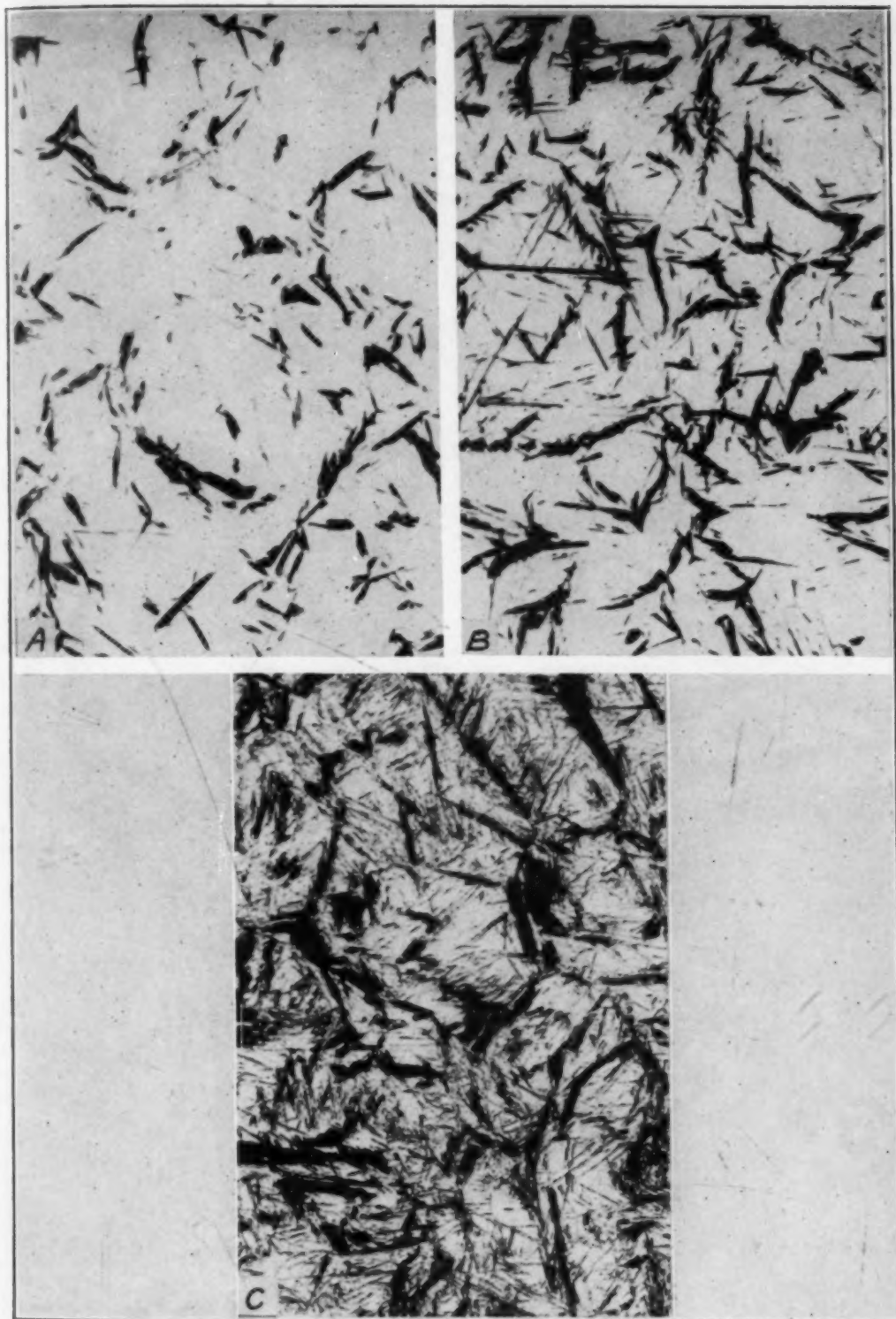


Fig. 3—0.82 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Nital. White Background Represents Austenite Not Decomposed at Temperature of Run. White Areas Contain a Mixture of Martensite and Austenite; the White Martensite is Formed During Final Quench to Room Temperature.

Fig. 3A—220 Degrees Cent.—1 Second. Fig. 3B—220 Degrees Cent.—3 Minutes. Fig. 3C—220 Degrees Cent.—15 Minutes.

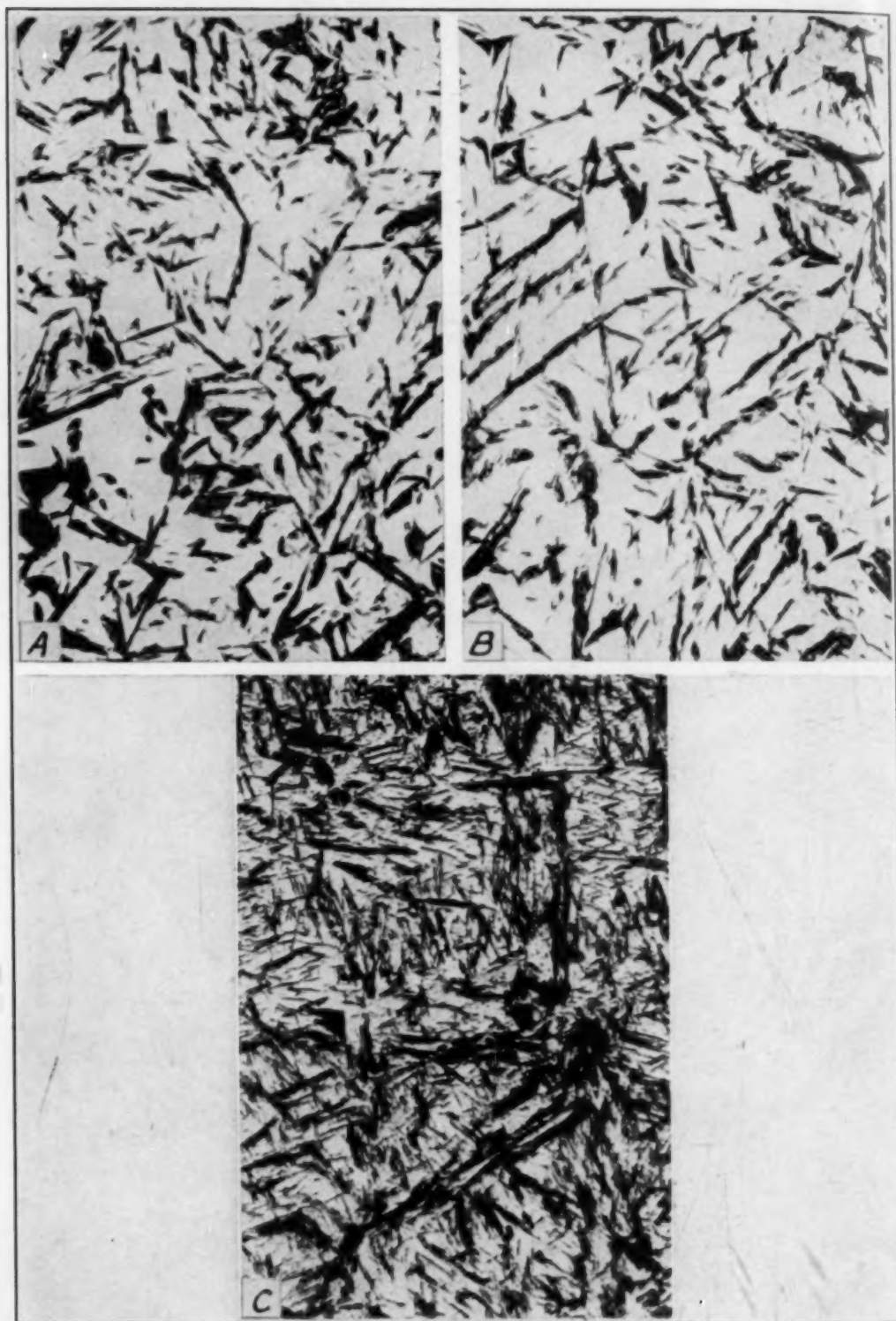


Fig. 4—0.82 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Nital. White Background Represents Austenite Not Decomposed at Temperature of Run. White Areas Contain a Mixture of Martensite and Austenite; the White Martensite is Formed During the Final Quench to Room Temperature.

Fig. 4A—200 Degrees Cent.—1 Second. Fig. 4B—200 Degrees Cent.—5 Minutes. Fig. 4C—200 Degrees Cent.—30 Minutes.

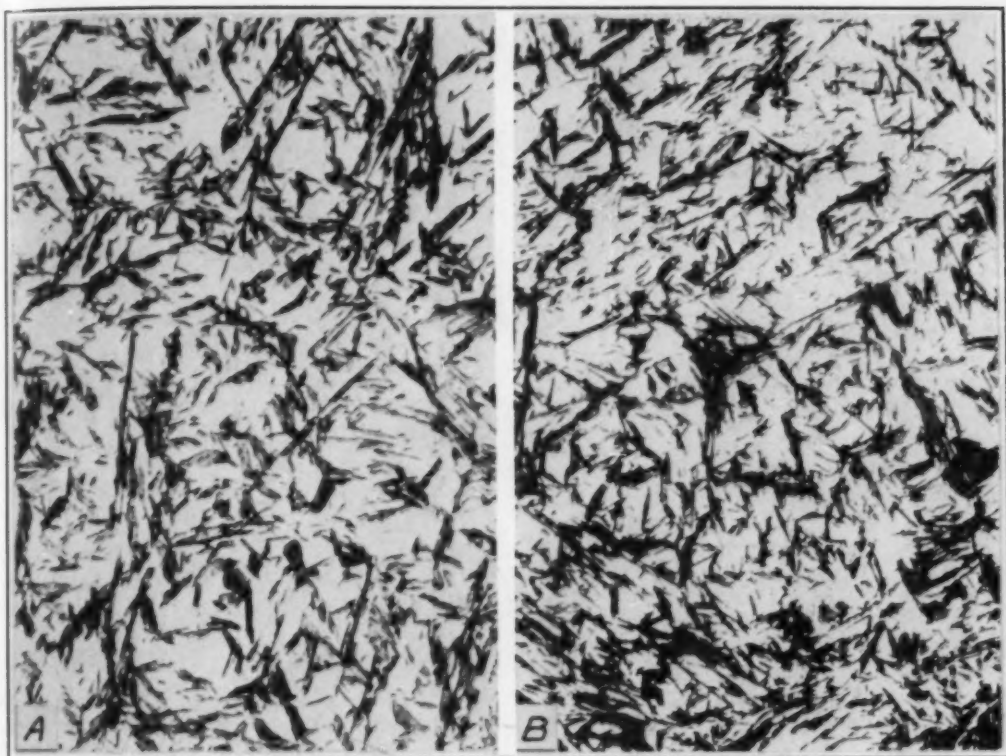


Fig. 5—0.82 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. 5 Per Cent Nital, 5 to 7 Seconds. $\times 500$.
Fig. 5A—180 Degrees Cent.—1 Second. Fig. 5B—180 Degrees Cent.—5 Minutes.

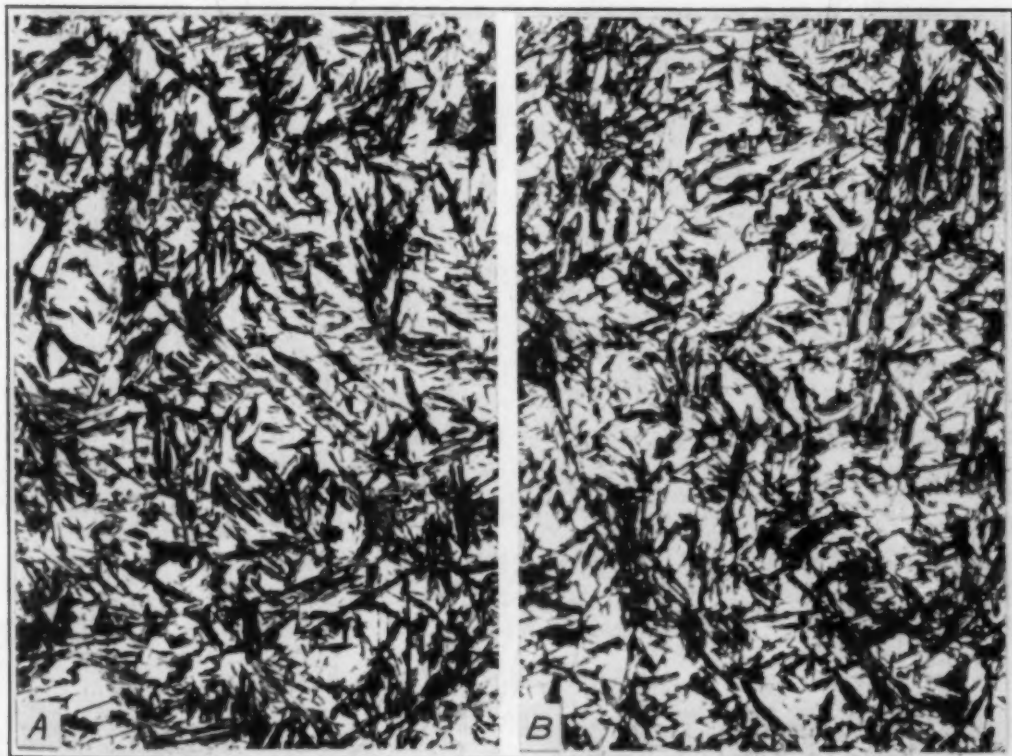


Fig. 6—0.82 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. 5 Per Cent Nital, 5 to 7 Seconds. $\times 500$.
Fig. 6A—160 Degrees Cent.—1 Second. Fig. 6B—160 Degrees Cent.—5 Minutes

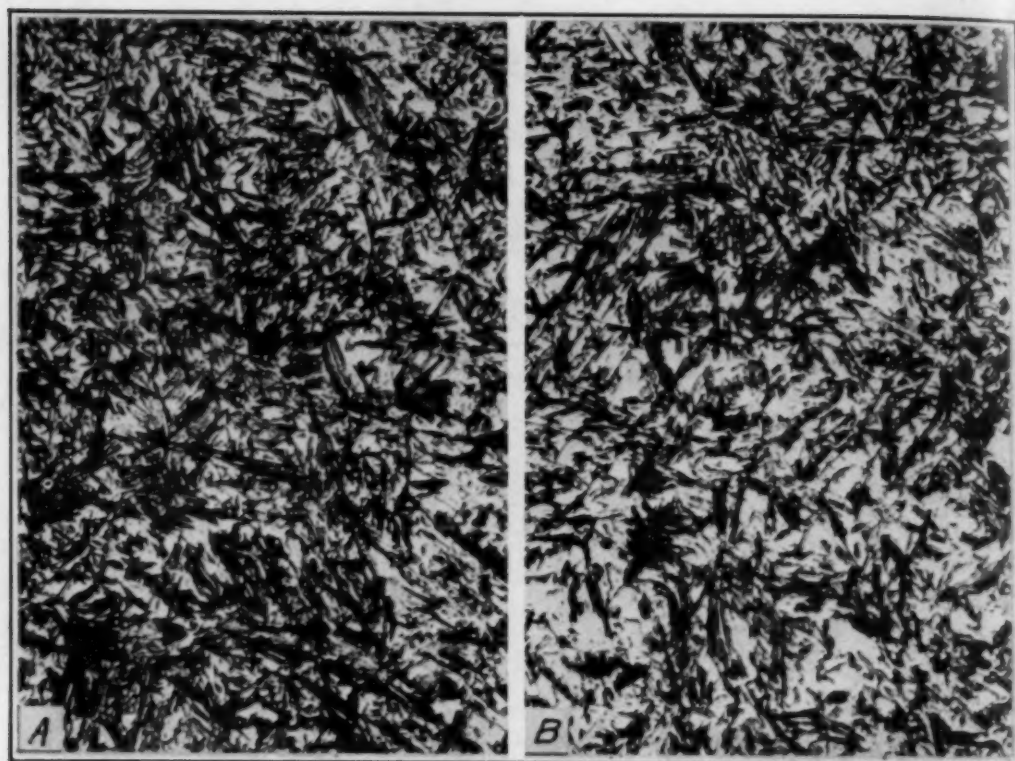


Fig. 7—0.82 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. See Text for Treatment Details. $\times 500$ Except as Noted. Etched 5 to 7 Seconds in 5 Per Cent Nital. In all Photomicrographs, White Background Represents Austenite That Had Not Decomposed at the Temperature of the Run. Actually These White Areas Contain a Mixture of Martensite and Austenite; the White Martensite is, of Course, Formed During the Final Quench to Room Temperature.

Fig. 7A—140 Degrees Cent.—1 Second. Fig. 7B—140 Degrees Cent.—5 Minutes.

nary eutectic mixture of Bi, Cd, Pb, and Sn; temperature controlled to within 1 degree Cent., while the tempering bath consisted of about five pounds of liquid tin maintained at a temperature of 275 to 280 degrees Cent. Final quench to room temperature in 10 per cent NaOH solution.

The quenching bath was located about 5 inches from the top outlet of furnace tube and the tempering bath was located about 4 inches from quenching bath. NaOH solution held within 2 or 3 inches from tempering bath for the final quench to room temperature.

Quenching Operation—Surfaces of each bath carefully skimmed before each quench; specimen rapidly removed from furnace and plunged into quenching bath for proper time; then rapidly plunged into tempering bath at 275 to 280 degrees Cent. for constant time of four to five seconds; then rapidly plunged into beaker of 10 per cent NaOH solution. Time for each transfer, furnace-to-bath or bath-to-bath, was only a fraction of a second.

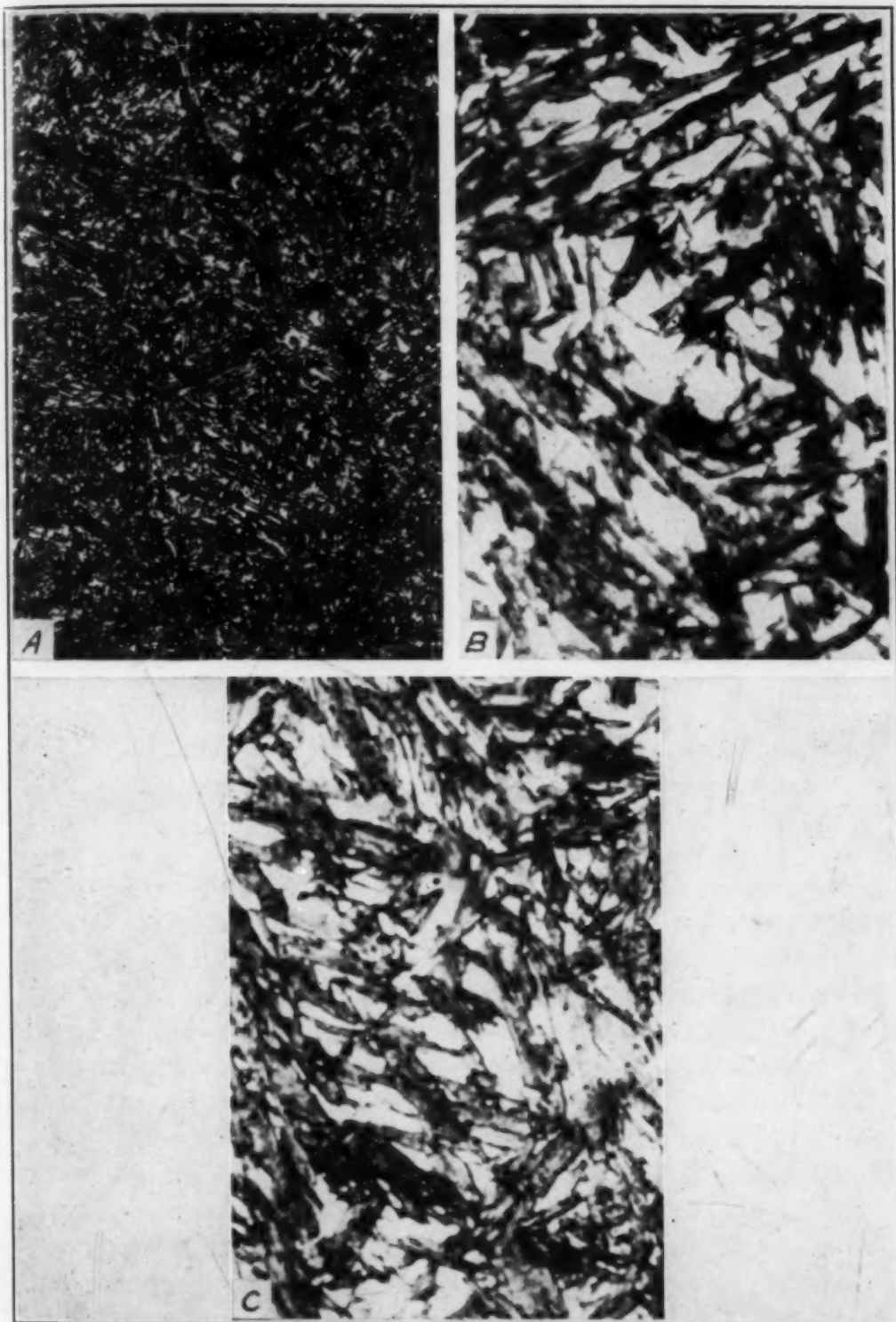


Fig. 8—0.82 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Nital.
Fig. 8A—100 Degrees Cent.—1 Second. Fig. 8B—100 Degrees Cent.—5 Seconds. $\times 2500$. Fig. 8C—100 Degrees Cent.—9 Days. $\times 2500$.

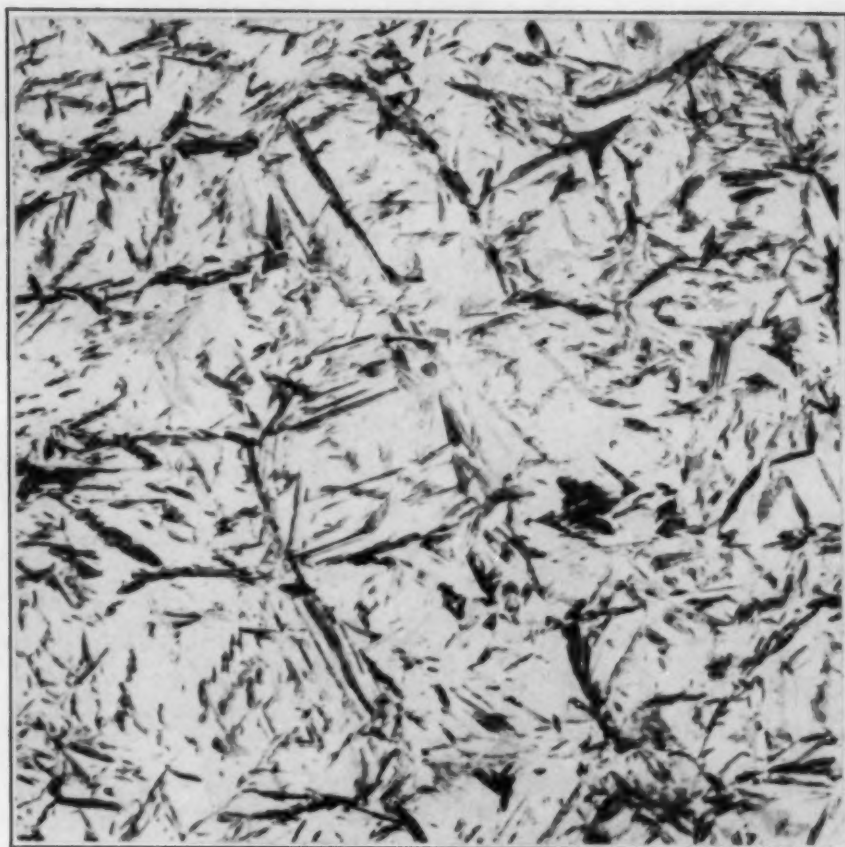


Fig. 9—0.82 Per Cent Carbon Steel Quenched to 200 Degrees Cent. for 2 Seconds, Then Quenched Directly to Room Temperature. $\times 500$. Etched in Nital.

Normal times of immersion for each temperature of quenching bath: 1 second, 2 seconds, 5 seconds, 30 seconds, 2 minutes, 3 minutes, 5 minutes, 10 (or more) minutes.

Small specimens only were used for the 1- and 2-second periods. Large specimens used for all periods of 5 seconds or more. For all quenching bath temperatures, one second was sufficient time for the entire volume of the small specimens to reach the bath temperature; for large specimens 5 seconds was sufficient, and 3 seconds was not. The following quenching bath temperatures were used: 260, 240, 220, 200, 180, 160, 140, 100 degrees Cent.

Microscopic Examination—The progress of transformation was followed microscopically. For polishing and photography the small specimens were held in clamps specially designed for the task. Surfaces of polish were located well below ($\frac{1}{8}$ inch or more) original specimen surfaces;¹⁰ and for small specimens, the surface of polish

¹⁰See Carpenter and Robertson (Ref. 3, p. 894) for a discussion of the surface effect in the decomposition of austenite.



Fig. 10—1.08 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. See Text for Treatment Details. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Absolute Nital. In all Photomicrographs, White Background Represents Austenite That Had Not Decomposed at the Temperature of the Run.

Fig. 10A—200 Degrees Cent.—75 Minutes. Fig. 10B—200 Degrees Cent.—6 Hours.

was a section passing through the center of volume. Photographs were taken near the centers of surfaces. All specimens were etched for a standard time of 5 to 7 seconds in 5 per cent absolute nital, and were photographed at magnifications of 500 or 2500 diameters. Estimations of per cent decomposition were made on photographic prints by the line-intercept method. Inaccuracy of estimation becomes greater the lower the temperature of the run.

RESULTS

The results are clearly shown in the accompanying photomicrographs. It should be noted that for this steel we are primarily concerned with the beginnings of decomposition, not with the end.

260 Degrees Cent.—Transformation did not begin until after about 2 minutes, and was about 30 to 40 per cent complete after 10 minutes.

240 Degrees Cent.—Transformation had started after 1 second (1 second is total time in bath; specimen was at temperature for

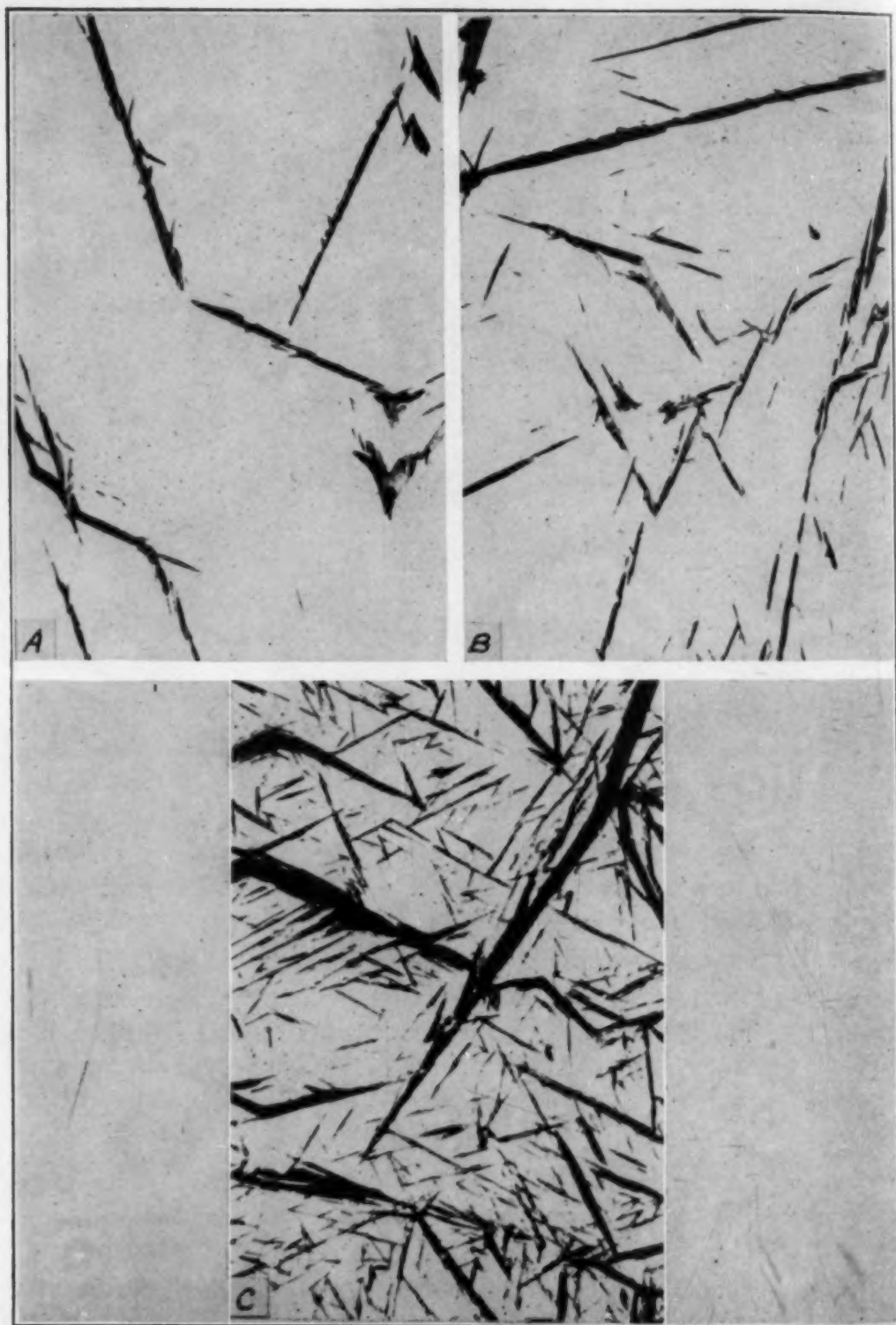


Fig. 11—1.08 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. See Text for Treatment Details. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Absolute Nital. In all Photomicrographs, White Background Represents Austenite That Had Not Decomposed at the Temperature of the Run.

Fig. 11A—160 Degrees Cent.—1 Second. Fig. 11B—160 Degrees Cent.—1 Hour.
Fig. 11C—160 Degrees Cent.—6 Hours.



Fig. 12—1.08 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. See Text for Treatment Details. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Absolute Nital. In all Photomicrographs, White Background Represents Austenite That Had Not Decomposed at the Temperature of the Run. 140 Degrees Cent.—1 Second.

only a fraction of a second). Transformation had produced very few martensite needles. The 1-second small specimen contained slightly more martensite than did the 5-second or 2-minute large specimens. No further decomposition was evident until at 3 minutes the beginnings of constant-temperature decomposition were observed. After 10 minutes the specimen was about 10 to 20 per cent decomposed.

220 Degrees Cent.—The 1-second specimen contained appreciable amounts of martensite, amounting to about 10 per cent of the total volume; the 2-second, 5-second, 30-second, and 2-minute specimens all showed about this same amount of martensite. At 3 minutes, constant-temperature decomposition was detected and after 15 minutes the specimen was about 20 to 30 per cent decomposed.

200 Degrees Cent.—For all times between 1 second and 5 minutes, decomposition was apparently the same. All specimens contained about 25 per cent martensite. After 30 minutes, the sum

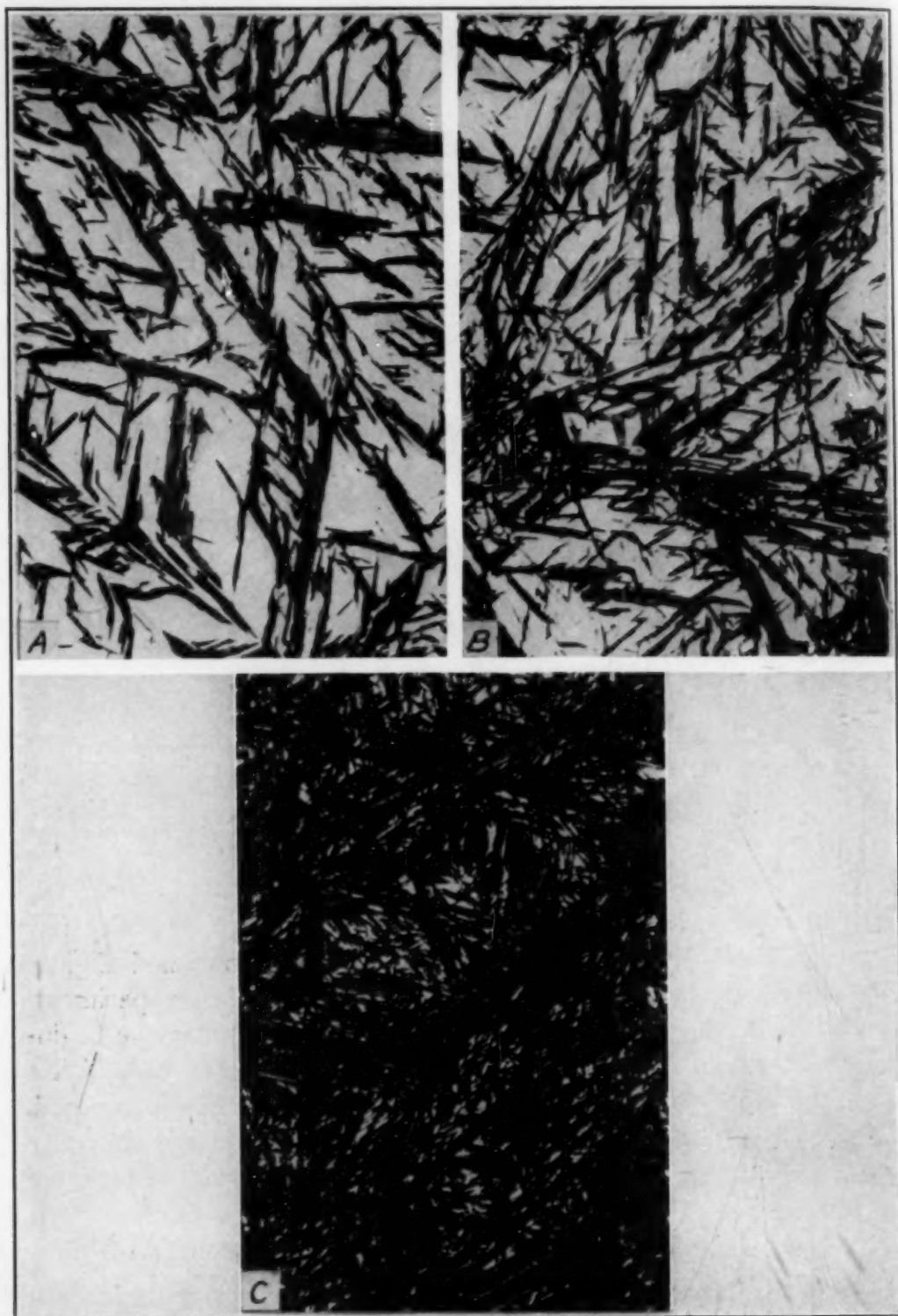


Fig 13—1.08 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. See Text for Treatment Details. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Absolute Nital. In all Photomicrographs, White Background Represents Austenite That Had Not Decomposed at the Temperature of the Run. Fig. 13A—120 Degrees Cent.—1 Second. Fig. 13B—120 Degrees Cent.—100 Hours. Fig. 13C—120 Degrees Cent.—600 Hours.

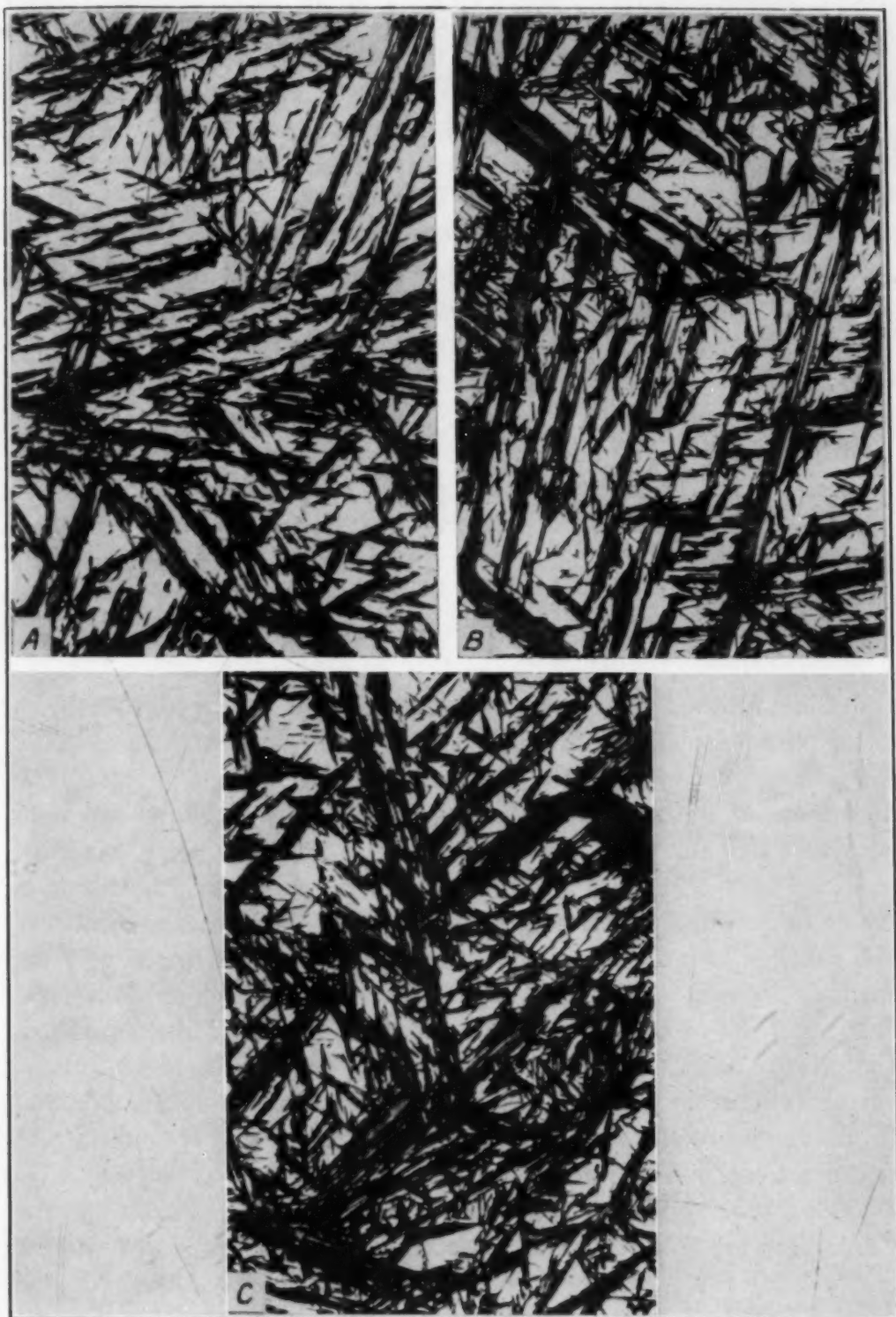


Fig. 14—1.08 Per Cent Carbon Steel Quenched to Various Temperatures and Held in Quenching Bath for Times Indicated. See Text for Treatment Details. $\times 500$. Etched 5 to 7 Seconds in 5 Per Cent Absolute Nital. In all Photomicrographs, White Background Represents Austenite That Had Not Decomposed at the Temperature of the Run. Fig. 14A—100 Degrees Cent.—1 Second. Fig. 14B—100 Degrees Cent.—240 Hours. Fig. 14C—100 Degrees Cent.—480 Hours.

of martensite plus the product of constant-temperature decomposition amounted to 30 to 35 per cent of the volume.

180 Degrees Cent.—For all times between 1 second and 5 minutes, so far as one could judge, decomposition was the same. Specimens contained about 35 per cent martensite. (Longer times were not used in the 180, 160, and 140-degree runs.)

160 Degrees Cent.—For all times between 1 second and 5 minutes, so far as one could judge, decomposition was the same. Specimens contained slightly more than 40 per cent martensite.

140 Degrees Cent.—For all times between 1 second and 5 minutes, so far as one could judge, decomposition was the same. Specimens contained slightly less than 50 per cent martensite.

100 Degrees Cent.—For all times between 1 second and 24 hours, so far as one could judge, decomposition was about the same. Specimens contained approximately 60 per cent martensite. After sixty days at temperature, decomposition was perceptibly greater.

Effect of Variables

The following factors might conceivably have affected the results reported above:

The 4-to-5-second tempering at 275 to 280 degrees Cent.—For most of the runs, representative short-time and long-time samples that had not received the tempering treatment were examined, and the conclusion was made that the tempering treatment had no effect on the amount of martensite produced at the temperature of the run.¹¹ An example is shown in Fig. 9, an "untempered" specimen after two seconds at 200 degrees Cent. (quenched directly from 200 degrees Cent. to room temperature). Thus, martensite that forms near 200 degrees Cent. or higher may be tempered to a certain extent in about a second near its formation temperature. In darkening-of-martensite effect, tempering for 5 seconds at 275 degrees Cent. is equivalent to about two days at 100 degrees Cent.

Austenite grain size—The temperature used (1200 degrees Cent.) for the solution anneal is considerably higher than that ordinarily used for heat treatment of eutectoid steel. This high temperature was used because the larger the austenite grain size the larger are the decomposition products for the temperature range

¹¹See Hanemann and Wiester (9) for photomicrographs proving this point. The proof is most completely made with alloys of high carbon content in which the austenite:martensite ratio is near 1:1 at room temperature.

studied. (In our work on very high carbon steel, we have found it expedient to use austenite grain sizes of 1 or 2 mm.). Grain-size variations were found to have little, if any, effect on the amount of martensite present in a specimen at a given temperature.

Impurities—A small specimen of S.A.E. 1090 steel (0.89 per cent carbon) after one second at 200 degrees Cent. contained slightly less tempered martensite than the equivalent specimen of the 0.82 per cent carbon relatively pure alloy used in the study. The slight difference is due mainly to the higher carbon content (see following pages).

DISCUSSION

The principal conclusions that may be drawn from the above results are as follows:

1. If specimens of eutectoid steel are hot-quenched to successively lower temperatures, the formation of martensite is first observed at a quenching-bath temperature of about 240 degrees Cent.
2. For temperatures below 240 degrees Cent. martensite is present a fraction of a second after the specimen has reached the quenching-bath temperature; the amount of martensite is greater the lower the temperature of the bath; and the austenite to martensite ratio at a given temperature does not change perceptibly until sufficient time has elapsed for diffusion processes to manifest themselves.
3. A reasonable assumption is that for this steel, martensite always begins to form, on quenching, at temperatures near 240 degrees Cent. and that martensite formation proceeds as the specimen temperature decreases.

The microscopic method of measuring reaction rates is obviously incapable of great accuracy when the decomposition product is so heterogeneous as it is for the temperature region here studied. Thus, from microscopic evidence alone, it is utterly impossible to state that, for example, at 200 degrees Cent. no decomposition occurred between times of one second and five minutes. However, it is certain that decomposition that may have occurred between one second and five minutes at 200 degrees Cent. is insignificant in comparison with the amount of martensite (25 per cent) that is present immediately after the specimen temperature has reached 200 degrees Cent. In contrast with these facts, the 1937 S-curve for eutectoid steel (5) states that at 200 degrees Cent. (390 degrees Fahr.) austenite does not begin to decompose until 2.5 minutes of time has elapsed. Simi-

larly, the S-curve states that decomposition at 140 degrees Cent. does not begin for 15 seconds, whereas the microscopic results show that a eutectoid specimen will contain about 50 per cent martensite by the time it has cooled to 140 degrees Cent. Obviously, the results of the present study of eutectoid steel substantiate at least qualitatively the older idea that martensite formation takes place only during cooling through the proper temperature range, and that martensite-formation temperatures cannot be depressed by increasing quenching velocity. It is not at all certain, however, that the same general rules will apply to any steel regardless of carbon content.

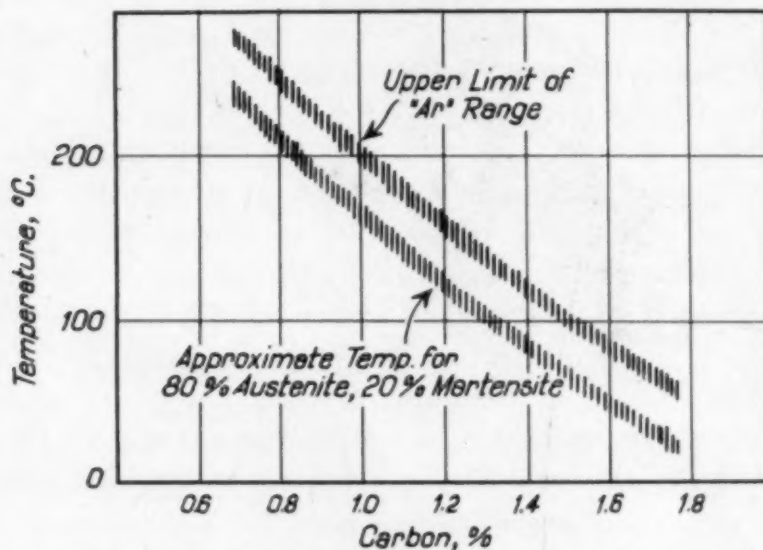


Fig. 15—Influence of Composition on Martensite Transformation Temperature Range for Plain Carbon Steel.

HYPEREUTECTOID STEELS

Studies similar to the one outlined above have also been made on hypereutectoid steel of the following carbon contents: 1.08 per cent carbon, 1.40 per cent carbon, and 1.78 per cent carbon. The conclusions made from studies of these high carbon steels are in general similar to those listed for eutectoid steel; however, inasmuch as the temperatures of the martensite range are lowered with increasing carbon content, detailed microscopic studies of the transformation are more readily carried out and thus a more complete picture of the transformation process can be given for hypereutectoid steels. On the other hand, studies of extremely high carbon steels are time-consuming, because of the extreme slowness of constant-temperature decomposition within or near the martensite range. For example,

the upper limit of the martensite range for 1.78 per cent carbon steel is about 65 degrees Cent. (see Fig. 15), and 5 to 10 per cent martensite may be produced by cooling this austenite from 65 to 45 degrees Cent., while on holding the specimen at this latter temperature, the first evidences of further decomposition are seen only after about four weeks at temperature.

The most complete study of the hypereutectoid series was made on a commercial drill rod steel containing 1.08 per cent carbon, and the results obtained from this steel are described below.

Experimental Methods—A $\frac{5}{8}$ -inch diameter rod was reduced by milling to a $\frac{1}{4}$ -inch square rod (to remove the major portion of the decarburized zone) and specimens were taken from this reduced section. The composition of the material was

C	Mn	Si	Cr	Ni
1.08	0.30	0.12	none	none

Except for temperatures and times, experimental methods were the same as those used for the study of eutectoid steel (see p. 13-16 above). All quenching was done from 950 degrees Cent. after a 10-minute solution anneal at 1200 degrees Cent. in an atmosphere of argon gas. Quenching-bath temperatures used were: 200, 180, 175, 170, 160, 140, 120, 100, and 80 degrees Cent.

Summary of Results

At 200 degrees Cent. the first traces of decomposition were observed after 1 hour at temperature. After 6 hours, decomposition was about 25 per cent, and was complete in 20 hours.

At 180 degrees Cent. the 1-second small specimen contained traces of martensite; in the 5-second, 5-minute, and 1-hour large specimens, very small amounts of martensite were present irregularly distributed at the surfaces only. The 1-hour specimen also contained the first traces of constant-temperature decomposition. After 6 hours, decomposition was about 10 per cent, and after 20 hours, about 90 per cent.

For lower quenching-bath temperatures, the amount of martensite present in the 1-second small specimens increased uniformly (see Figs. 10-14); and at least for the 170 and 160 degrees Cent. runs the time required to detect the first constant-temperature decomposition was noticeably less than for the 180 and 200 degrees Cent.

runs.¹² It is possible that additional decomposition was present even after 1 minute at the former temperatures; it is certain that both the 170 and 160 degrees Cent. 1-hour specimens had suffered more constant-temperature decomposition than had the 1-hour specimens at 180 and 200 degrees Cent. For times of 6 hours or longer the amount of constant-temperature decomposition (disregarding the martensite initially present) was always considerably less the lower the temperature of the run; for the 100 degrees Cent. run, specimens contained 35 to 40 per cent martensite immediately after reaching bath temperature, and constant-temperature decomposition could not be seen until after ten days at temperature.

HYPOEUTECTOID STEEL

An attempt was made to apply the microscopic method to the study of martensite formation in S.A.E. 1050 steel. The method was found to be unsuited to the study of this steel for the following reasons:

1. Although it is practically certain that martensite formation in 0.50 per cent carbon steel occurs continuously during cooling (for any rapid cooling velocity) beginning at about 300 degrees Cent., there appears to be no readily measurable lag between the time the specimen reaches temperature (near 300 degrees Cent.) and the time for the appearance of constant-temperature decomposition products.
2. The microscopic method of study hinges upon the ability to differentiate between dark (tempered) martensite and white martensite, and this is easily done when the major portion of the temperature range being studied is below about 200 degrees Cent. For low carbon steels, however, martensite forms at a relatively high temperature and a certain amount of tempering always occurs during quenching. It is undoubtedly for this reason that a rapidly quenched 0.30 per cent carbon steel will always show more structure detail (variations in light and dark shadings of martensite crystals) than will a similarly quenched 0.70 per cent carbon steel. In the former, the martensite that forms near 350 degrees Cent. will be tempered during the quenching operation considerably more than will the portion that forms at, say, 200 degrees Cent. In the latter steel, martensite will form only at temperatures below about 250 degrees Cent.,

¹²The fact that constant-temperature decomposition is *initially* more rapid below the upper limit of the *Ar''* range than above was established also for the 1.4 and 1.78 per cent carbon steels.

and it is possible to quench through this range rapidly enough to avoid any noticeable tempering during the quench.

CONCLUSIONS

In general, the interpretation that Carpenter and Robertson have given to the austenite-martensite transformation has been confirmed in this study. However, certain modifications or refinements should be stressed. In the first place, it is necessary to distinguish between the martensite transformation in very low carbon steels and that in eutectoid and hypereutectoid steels. Obviously, the martensite process cannot be considered completely diffusionless (as it *must* be in its pure form, that is, near or below room temperature) when it occurs at a sufficiently high temperature for diffusion processes to become a complicating factor. Thus, it appears best to restrict this summary to eutectoid and hypereutectoid steels, admitting that in low carbon steels the martensite process may be either fundamentally different or merely different in degree.

For heat treating techniques making use of the interrupted quench (that is, for the "austempering" treatment), the temperatures for the beginning of martensite formation as a function of carbon content are shown in Fig. 15.¹³ These temperatures are insensitive to the normal impurities of commercial plain carbon steels.¹⁴ Time is not a significant variable for the process represented in Fig. 15; that is, the martensite transformation cannot be depressed to lower temperatures by increasing the cooling velocity.

If a small specimen is quenched into a liquid metal bath held at a temperature within the martensite range for the specimen, the decomposition process may be subdivided as follows:

1. Martensite will begin to form as soon as the specimen temperature reaches the upper limit of its martensite range.
2. Martensite will continue to form as the cooling proceeds.
3. A slight amount of martensite may be formed after the

¹³This curve agrees well with the results of Hanemann and Wiester (9), who in 1931 reported a microscopic determination of A_r temperatures for plain carbon steels of commercial purity. Hanemann, in his researches, was so impressed with the comparative stability of the austenite:martensite ratio at temperatures below 200 degrees Cent. that he dismissed time as a variable and regarded martensite as an equilibrium phase, which, of course, it is not.

¹⁴For a given carbon content, the temperature of the martensite range is depressed by both nickel and manganese; manganese is considerably more effective than nickel in this lowering. For example, in Hadfield steel (about 12 per cent manganese and 1 per cent carbon) the upper limit of A_r range is near -200 degrees Cent. In an alloy of 12 per cent nickel and 1.1 per cent carbon, the upper limit of this range is at room temperature. In high alloy steels, these temperatures are perceptibly affected by cooling stresses and, within limits, martensite ranges are raised as cooling velocities are increased.

specimen has attained temperature equilibrium. This is a minor effect, however, in comparison with the martensite formed during cooling.

4. As the specimen is maintained at constant temperature, the remaining austenite will decompose. This decomposition will proceed according to some complex function of time and temperature. However, for appreciable amounts of decomposition (5 per cent or more), the rate of constant-temperature decomposition of the remaining austenite in a given steel will always be very much slower the lower the temperature of the quenching bath.¹⁵ It appears certain that the rate of constant-temperature decomposition is *initially* greater for temperatures just below the upper limit of the martensite range than for temperatures just above this range. However, it will be impossible to evaluate the importance of this effect until quantitative reaction-rate curves are evaluated by some method capable of measuring both martensite formation during rapid quenching and subsequent decomposition as a function of time after the specimen has reached temperature equilibrium with the quenching bath.

Again, it should be emphasized that the microscopic method used in this study is incapable of high accuracy; however, it unquestionably gives the most reliable direct evidence survey of the entire process of austenite decomposition, and thus should serve as a basis for the application of more refined and quantitative methods of study.

It is concluded that the S-curve of austenite decomposition does not give a correct picture of the austenite to martensite transformation in steel.

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¹⁵In our discussion of austenite decomposition at constant temperature, we have found it helpful to accept the following working hypothesis: At temperatures within or near the martensite range, the bulk of austenite decomposition at constant temperature has its start in the precipitation of thin sheets of cementite (or a carbon-rich transition lattice), which would necessarily be accompanied by a depletion in carbon content of the austenite surrounding each cementite sheet. This lowering of carbon content of austenite regions by the precipitation of thin, perhaps invisible, cementite sheets would effectively "bring the martensite range" to these regions of the specimen, and would permit the formation of pseudomorphic martensite about each cementite sheet while the temperature is being held constant. The main reason for adopting this hypothesis is that the formation of martensite pseudomorphic to cementite actually does take place in high carbon steels with certain types of heat treatment. Proof of this will be given in another publication devoted to the crystallography of austenite decomposition.

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DISCUSSION

Written Discussion: By Morris Cohen, assistant professor of physical metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

To those who have been closely following the literature published in this country and abroad on the subject of subcritical transformations in steel, it has become increasingly obvious that the American metallurgists and the European metallurgists are on opposite sides of the ocean, not only geographically speaking, but also martensitically speaking. When we talk about martensite formation in this country, we usually have in mind a family of S-curves which indicate that martensite may form as a function of time and of temperature; but abroad the feeling is that time has little or nothing to do with martensite formation: namely, that martensite begins to form when a certain low temperature is reached, and continues to form during further cooling because the temperature is changing. The elimination of time as a factor in this transformation is consistent with the concept that martensite formation does not have to depend upon a diffusion process. We may readily visualize, therefore, that such a transformation would be difficult, if not impossible, to suppress by rapid cooling.

Our work on the retained austenite in high speed steel¹⁰ has led to similar conclusions. When quenched high speed steel is tempered, the retained austenite does not transform at the tempering temperature, but during cooling from the tempering temperature. The temperature at which this transformation begins during cooling and the hardness of the martensite thus formed are substantially independent of the rate of cooling. If the cooling from the tempering temperature is stopped during this austenite-martensite transformation (in other words, within the A_r'' temperature range), the transformation quickly stops. On continued holding at constant temperature within the A_r'' range, the austenite eventually begins to transform isothermally. The nature of this constant temperature product is unknown, but it is too soft to be martensite or even martensite which has been tempered at the holding temperature. If the cooling is started again before all of the austenite is isothermally transformed, there is virtually no transformation during cooling through an appreciable drop in temperature, after which the remaining austenite starts to transform into martensite. The greater the extent of the isothermal decomposition at the holding temperature, the greater is the temperature lag during subsequent cooling before the austenite-martensite reaction sets in again. Thus, we see that the kinetics of the residual austenite transformation in high speed steel are similar to the kinetics of the primary austenite transformation in plain carbon steels.

Inasmuch as the S-curve representation of austenite transformations at subcritical temperatures has proved so useful in our understanding of the structural changes which occur in steel, it is natural to inquire whether the lower part of the S-curve diagram can be modified so as to emphasize the temperature factor and minimize the time factor in the austenite-martensite reaction. At the same time, the modified diagram should show that even though the martensite formation stops when the cooling is stopped, the remaining austenite ultimately undergoes isothermal decomposition into a product which is probably not the same as the martensite which forms during cooling.

Such a diagram has been devised by the writer, and is shown in Fig. A. It should be emphasized that this diagram is intended only as a qualitative representation, and is not to be construed as an actual plot. It is offered simply as a schematic picture which may serve as a connecting link between the two divergent schools of thought on the subject of martensite formation. For the sake of simplicity, it is assumed in this diagram that the isothermal transformation product formed at moderately low temperatures falls in the same class as the bainite structures. Hence, the term "bainite" is used in a loose sense to cover all of the isothermal transformation products other than pearlite. (The essence of this discussion would not be changed if we assumed the existence of several kinds of bainite.) Unlike martensite, the formation of these products depends upon diffusion and hence may require appreciable time.

The upper part of the S-curves remains unchanged. At temperature above the nose of the curves the austenite transforms isothermally into pearlite or lamellar structures, and at lower temperatures the austenite transforms isothermally into bainite or acicular structures. In general, the rate of bainite

¹⁰M. Cohen and P. K. Koh, "The Tempering of High Speed Steel," *TRANSACTIONS, American Society for Metals*, Vol. 27, 1939, p. 1015.

formation decreases with decreasing temperature and the corresponding reaction curves are C-shaped rather than S-shaped. In the lower part of the diagram, a range of temperatures is indicated as the Ar'' range. If austenite is cooled through this temperature range, it transforms into martensite. This austenite-martensite transformation starts when the 0 per cent horizontal (beginning of the Ar'' range) is reached and continues during cooling so that the trans-

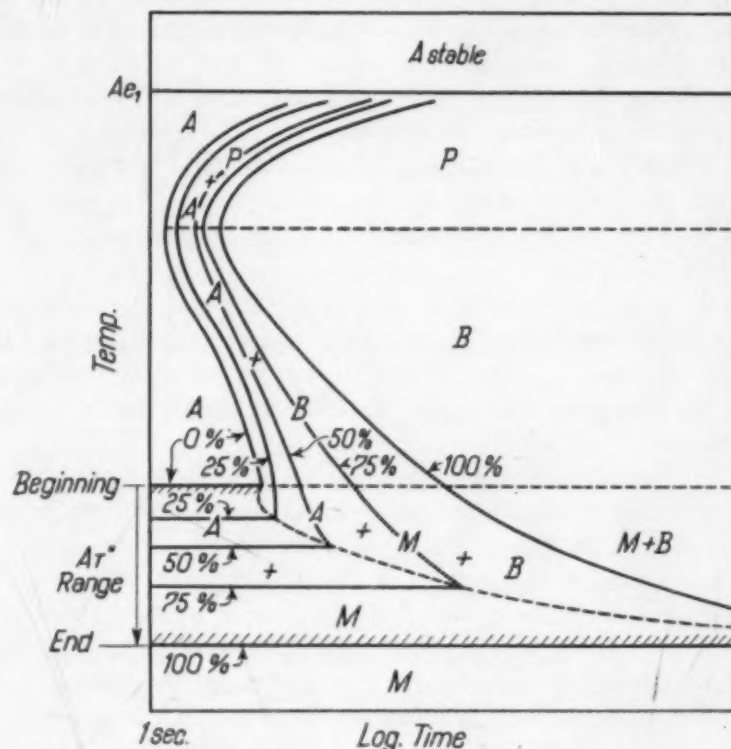


Fig. A—Modified S-Curve Diagram. Full Lines Represent Per Cent of Austenite Transformed. A = Austenite, P = Pearlite, B = Bainite, M = Martensite or Tempered Martensite.

formation is 25 per cent complete when the temperature of the 25 per cent horizontal is reached and is 50 per cent complete at the 50 per cent horizontal, and so on. Hence, within the Ar'' range, the temperature to which the austenite is cooled controls the extent of the austenite-martensite transformation, and the rate of cooling in this region has no detectable influence on this transformation.

This diagram also shows what happens when austenite is cooled to some temperature within the Ar'' range and is then held at this constant temperature. For example, if this temperature corresponds to the 50 per cent horizontal, 50 per cent of the austenite will have transformed into martensite by the time the holding temperature is reached. At this temperature, little if any austenite transforms isothermally until the time where the 50 per cent horizontal intersects the 50 per cent isothermal C-curve. From this time on, the austenite decomposes isothermally (into bainite, we assume) until the austenite completely disappears. In accordance with experimental facts, the 100 per cent C-curve is drawn so that it will not intersect the 100 per cent horizontal for an

exceedingly long time. Hence, if room temperature happens to lie just above the lower limit of the Ar'' range, the small amount of austenite which does not transform into martensite during the cooling to room temperature will remain in the steel for an almost indefinite length of time without undergoing isothermal transformation.

In this diagram, the dotted line which represents the locus of the intersections of the isothermal C-curves with the corresponding Ar'' temperature horizontals is purposely drawn to denote a slightly faster onset of the isothermal transformation at a temperature just within the Ar'' range than at a temperature just above the Ar'' range. This phenomenon, which is emphasized by the authors in the footnote on page 560, has been noted independently by the writer, and therefore has been incorporated in the schematic diagram. Possibly the stresses caused by the formation of the voluminous martensite during the cooling to the holding temperature hasten the initial stages of the subsequent isothermal transformation.

It is also interesting to note that this diagram is consistent with the experimental fact that the greater the extent of isothermal transformation which takes place at any given holding temperature within the Ar'' range, the greater is the temperature lag during subsequent cooling before the remaining austenite begins to transform into martensite.

When we compare this schematic diagram with the original set of S-curves (Fig. B) published by Davenport and Bain,¹⁷ we find a surprising degree of similarity. The promontories in the S-curves do not occur at the same temperature, (as is the case at the nose of the curves), but appear at successively lower temperatures as we compare each S-curve with the next. Furthermore, the lower part of each S-curve is practically horizontal. The schematic diagram in Fig. A shows these same characteristics, and suggests that the S-curves may represent the experimental determination of a family of isothermal C-curves which are intercepted at lower temperatures by a family of Ar'' temperature horizontals due to the setting in of the rapid martensitic reaction which occurs during the cooling to the holding temperature within the Ar'' range.

On this basis, the bending-back of the S-curves in Fig. B to form the lower side of the promontories at temperatures below approximately 250 degrees Cent. should not be interpreted as an increase in the rate of isothermal transformation into bainite, but simply that these temperatures come within the Ar'' range, and hence the rapid austenite-martensite reaction sets while the steel is still cooling to the holding temperature.

The only discrepancy between the schematic diagram in Fig. A and the experimental curves in Fig. B lies in the tail ends of the S-curves at very low temperatures. Since these tail ends all come within a 5-second time interval, and since it is doubtful that the dilatometer specimens used for this experimental determination could actually reach the low holding temperature within 5 seconds, the experimental error in ascertaining the true "zero" time at the holding temperature may be of the same order of magnitude as the time intervals plotted. Therefore, too much weight should not be placed upon these tail ends. This is

¹⁷E. S. Davenport and E. C. Bain, "Transformation of Austenite at Constant Subcritical Temperatures," *Transactions, American Institute of Mining and Metallurgical Engineers*, Vol. 90, 1930, p. 117.

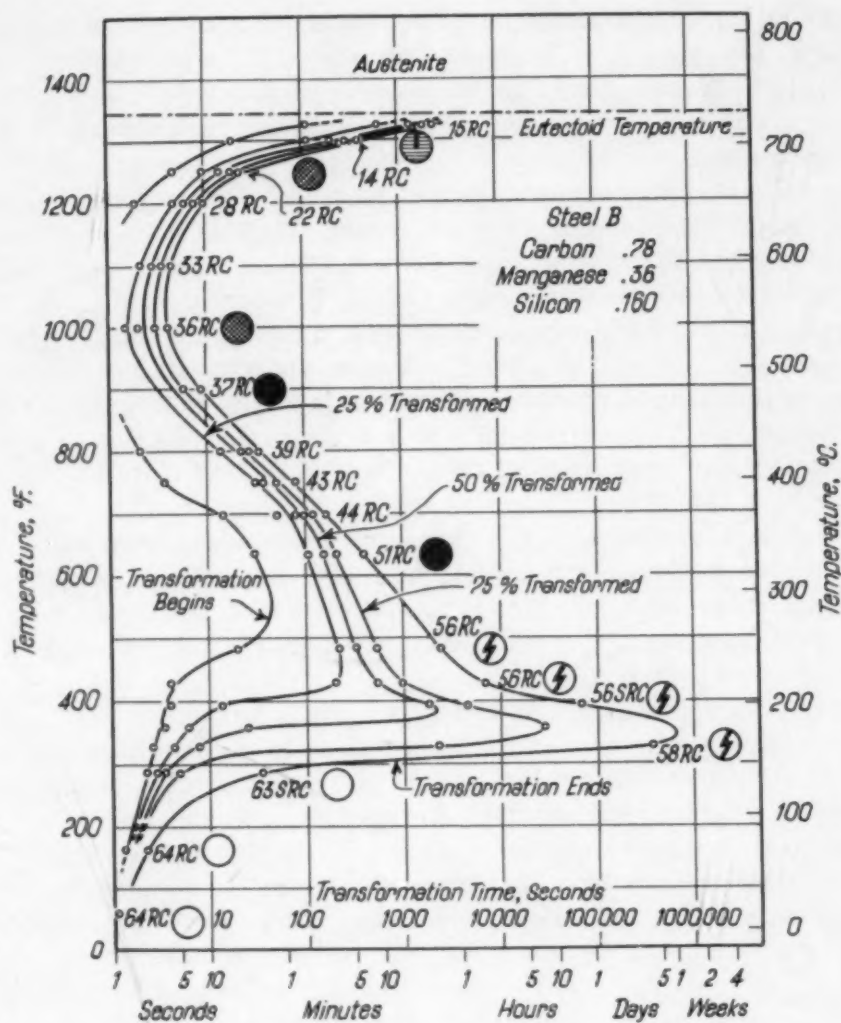


Fig. B—Original Set of S-Curves for "Eutectoid" Steel Published by Davenport and Bain.

confirmed by an inspection of the dilatometer curves published by Davenport and Bain.¹⁸ These curves show that the lower the holding temperature, the less is the total expansion which is observed. But just the opposite should be true if all the austenite could be made to transform isothermally in this temperature range because the low temperature products have relatively large specific volumes. Hence, we may infer that in these dilatometer experiments some of the austenite transformed during the cooling to the holding temperature, and consequently the lower the holding temperature, the less was the amount of austenite which remained to transform isothermally. It is also conceivable that at least a part of the expansion shown by the dilatometer curves at the lower temperatures did not occur at the holding temperatures but rather during the cooling through the last few degrees before the holding temperature was actually reached.

Thus, because of the experimental difficulties involved, there appears to be

¹⁸Transactions, American Institute of Mining and Metallurgical Engineers, Vol. 90, 1930, p. 117.

some justification for disregarding the tail ends of the S-curves. If this is done, the S-curves are found to be in good agreement with the schematic diagram in Fig. A, which in turn seems to be quite consistent with the ideas of the European metallurgists. The present writer feels that this schematic diagram offers a common ground upon which investigators with different points of view on the question of the austenite-martensite transformation may possibly get together to ascertain the cause of their differences.

Written Discussion: By Howard Scott, section engineer, metallurgical section, Westinghouse Electric & Mfg. Co., E. Pittsburgh, Pa.

The mechanism of austenite decomposition in carbon steels at low temperatures is a highly controversial subject. A clear understanding of this reaction in fact seems remote unless a conclusive answer is given to the question: Is the temperature at which Ar'' starts in carbon steels so quenched as to retain all carbon in austenitic solid solution independent of cooling rate at rates faster than the critical value?

We are quite familiar with this transformation in low carbon alloys where it is called Ar_s and know that cooling rate with them is a negligible factor. The authors believe that the same statement applies to carbon steels. Their data, however, supports that proposition better than their argument, hence the present amplifying remarks.

Direct evidence for the affirmative is difficult to obtain. Perhaps the best direct evidence that can be cited is the fact that the authors have determined the start of Ar'' in an 0.82 per cent carbon high purity steel to be 250 degrees Cent. with an accuracy close to ± 10 degrees Cent. Digges, in another paper on this program, determined the same property in an 0.80 per cent carbon pure iron-carbon alloy by an entirely different method to be 266 degrees Cent. Scott and Hoop¹⁹ by still another method, give values whose mean is close to those of Digges over the range 0.4 to 0.8 per cent carbon, as shown in the writer's discussion of Digges' paper. It is, therefore, well established that determinations of this transformation though difficult are reproducible with an accuracy which is good under the circumstances. Such agreement is possible only if cooling rate is a minor factor.

Despite this conclusion the popular contrary view has a material factual basis. How the physical observations involved can be easily interpreted one way or the other may be seen by assuming that Ar'' is unaffected by variation in cooling rates above the critical value, and then noting the reactions to be expected. Incidentally, the fact that the effects predicted from this assumption are just those observed is perhaps the best support available for it. Accordingly, the authors' observations are examined in the following paragraphs on this basis.

In their first experiment the authors quenched an 0.82 per cent carbon steel into a bath at 260 degrees Cent., about 10 degrees Cent. above the start of Ar'' . This transformation, therefore, should not start at this temperature but actually some reaction does. After 2 minutes a small amount of a dark constituent is observed, Fig. 2A, an obvious contradiction to our assumption. The contradiction is resolved immediately, however, when it is realized that at this tem-

¹⁹Scott and Hoop, TRANSACTIONS, American Society for Metals, Vol. 22, p. 233.

perature the austenite is highly supersaturated with respect to carbon as may be seen by extrapolating the cementite solid solubility line of the iron-carbon diagram to the same temperature. Being still at a temperature where carbon mobility is appreciable, cementite needles precipitate, Fig. 2B and 2C. The austenite adjacent to the cementite needles is impoverished in solid solution carbon so martensite forms there as required by the fact that the start of A_r'' is higher the lower the solid solution carbon content. Thus one may explain the confusing fact that a martensite constituent forms at a temperature higher than normal for the martensite transformation.

The constituent formed secondary to carbide precipitation is easily distinguished both by its structural appearance and by the rate at which it forms as may be seen by comparing Figs. 2 and 3. The latter figure represents specimens quenched to 220 degrees Cent. which is 30 degrees Cent. below the start of A_r'' . Here considerable primary martensite forms in one second while three minutes at 260 degrees Cent. does not produce as much of the secondary constituent. This constituent also has a distinctly different appearance from that formed at 260 degrees Cent. Fine needles appear prominently at 220 degrees Cent. only after holding 15 minutes, Fig. 3C, and they are finer than when formed at 260 degrees Cent. as would be expected from the known loss in mobility of carbon with drop in temperature. Also there is scarcely more of the coarse martensite formed by direct transformation at 220 degrees Cent. after 15 minutes' exposure than after one second, which observation conforms with the initial assumption.

Similar experiments on a 1.08 per cent carbon steel gave an even more striking confirmation of this view. Figs. 10 and 11, for example, represent specimens quenched respectively 20 degrees Cent. above and 20 degrees Cent. below the start of A_r'' . At the upper temperature no primary martensite forms and the secondary constituent appears only after 75 minutes' exposure. Even 6 hours at the lower temperature does not increase greatly the amount of primary martensite over that formed in one second.

These observations show as has not been done previously how extremely sharp is the transition from the secondary reaction to direct formation of martensite. This fact is not indicated in current models of reaction rate curves so we must agree with the authors that these curves need revision at the lower end. Also there can now be little doubt that precipitation of carbide is the reaction determining the form of the reaction rate curve in the region between its peak and the start of A_r'' .

Written Discussion: By J. R. Vilella, research laboratories, United States Steel Corp., Kearny, N. J.

I wish to call your attention to several points in this paper which were not emphasized during its oral presentation. The authors' main conclusion is that the S-curve of austenite decomposition does not give a correct picture of the austenite to martensite transformation in steel. To arrive at this conclusion the authors investigated the transformation of austenite at temperatures below 500 degrees Fahr. (260 degrees Cent.), employing in their work an iron-carbon alloy of far greater purity than the commercial steels used by Davenport and Bain. Since I have not investigated the isothermal transformation of such high purity iron-carbon alloys I am not commenting on the data presented in this

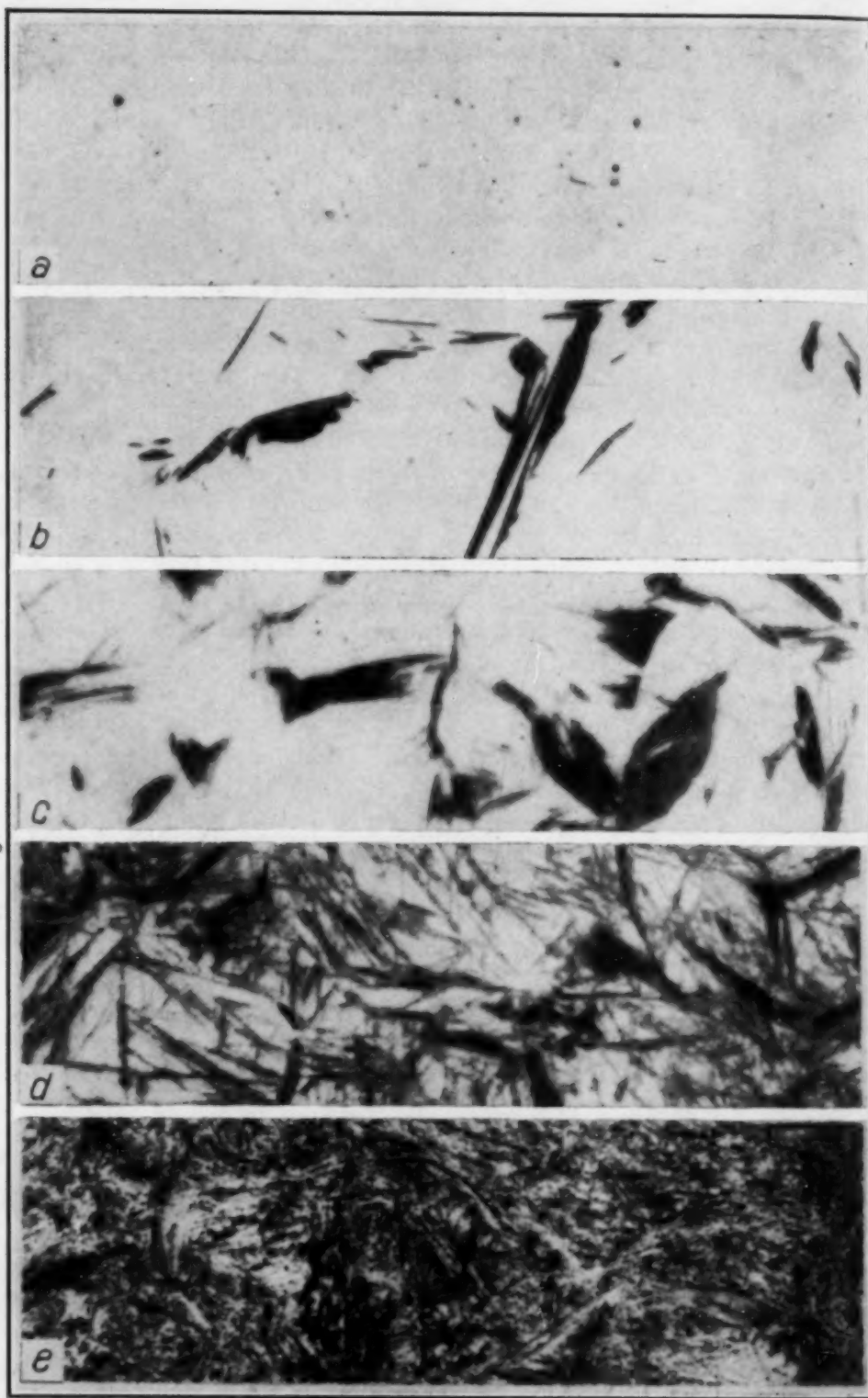


Fig. 1—Showing Progress of Transformation at 400 Degrees Fahr. (240 Degrees Cent.). Fig. 1a—Transformation Time: 2 Seconds. Fig. 1b—Transformation Time: 30 Seconds. Fig. 1c—Transformation Time: 16 Minutes. Fig. 1d—Transformation Time: 1 Hour. Fig. 1e—Transformation Time: 20 Hours.

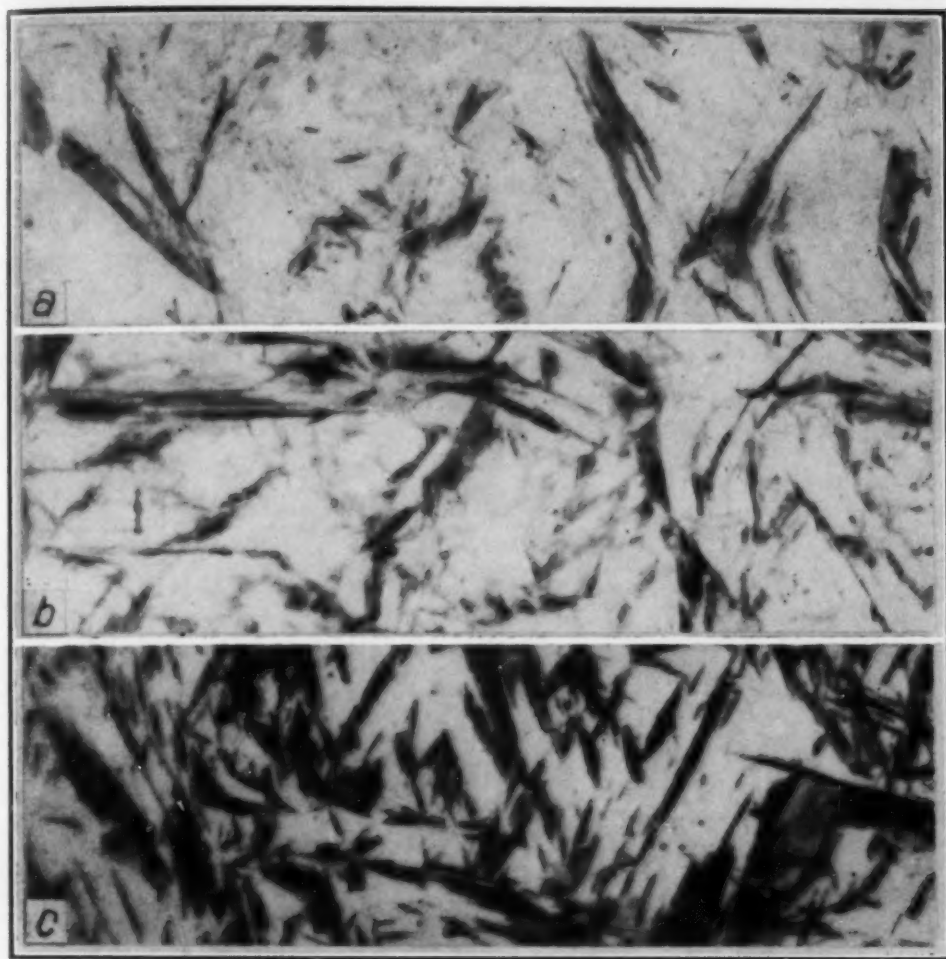


Fig. 2—Showing Progress of Transformation at 300 Degrees Fahr. (149 Degrees Cent.). Fig. 2a—Transformation Time: 2 Seconds. Fig. 2b—Transformation Time: 15 Seconds. Fig. 2c—Transformation Time: 60 Seconds.

paper, but I wish to present some results obtained since reading a preprint of this article, which are at variance with the authors' conclusions. A commercially made steel of the following composition was used: carbon 0.89 per cent; manganese 0.29 per cent; silicon 0.15 per cent; chromium 0.11 per cent.

Fig. 1 illustrates the austenite-martensite transformation at 400 degrees Fahr. (204 degrees Cent.). According to Greninger and Troiano a eutectoid steel at the end of 2 seconds should be approximately 25 per cent transformed to martensite and should continue unchanged for at least 5 minutes. Actually at the end of 2 seconds no transformation has taken place and at the end of 16 minutes the transformation is approximately 10 per cent completed. It is evident, therefore, that the data obtained in high purity iron-carbon alloys are not necessarily correct for commercial steel of similar carbon content. On the strength of these data it must be concluded that the S-curve gives a truer picture of the transformation of this steel at 400 degrees Fahr. than can be drawn from the authors' data.

Fig. 2 illustrates the transformation at 300 degrees Fahr. (149 degrees

Cent.). According to Greninger and Troiano, at the end of 1 second the steel should be nearly 50 per cent transformed to martensite and should remain unchanged at least up to 5 minutes, but as may be seen in the photomicrographs, at the end of 2 seconds only approximately 5 per cent "transformation product" is present (large, darkly etched needles) and with increasing time at temperature the transformation advances constantly in the manner indicated by the S-curve.

In the various S-curves published by my associates it has been pointed out

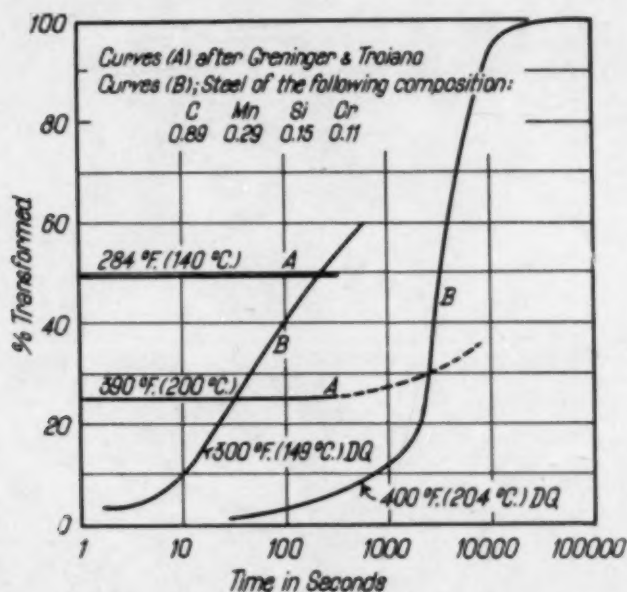


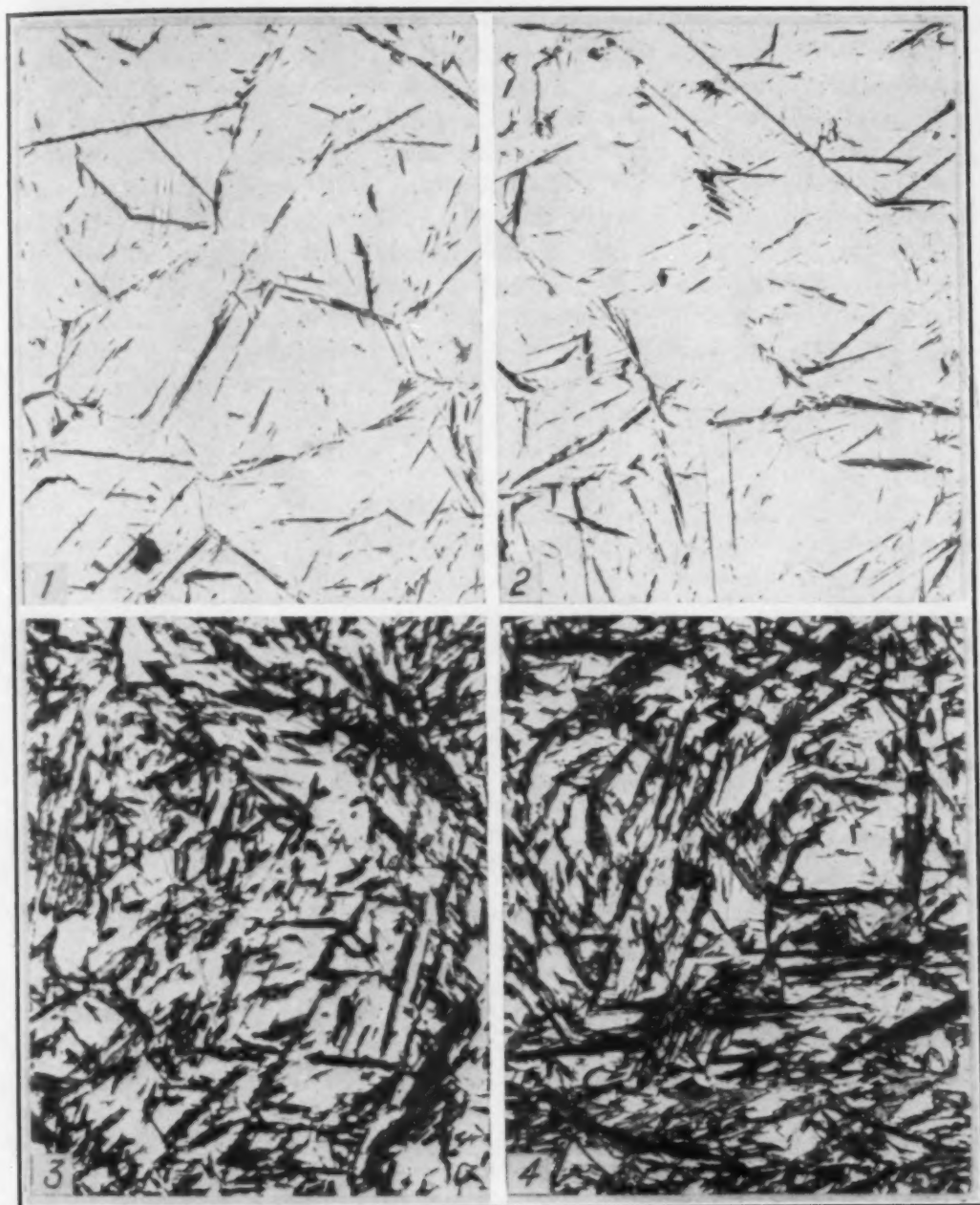
Fig. 3—Curves A are a Graphic Representation of How the Transformation Should Proceed According to Greninger and Troiano. Curves B are Drawn from Microscopic Estimation of the Amount of Transformation Actually Observed in the Commercial Steel Examined.

that the temperature zone of least accuracy is that lying below 300 degrees Fahr., because herein lie the greatest experimental difficulties. The data cited in this discussion, however, are not subject to this criticism since they are clear cut and accurate.

Curves marked A in Fig. 3 are a graphic representation of how the transformation should proceed according to Greninger and Troiano and curves marked B those drawn from the microscopic estimation of the amount of transformation actually observed in the commercial steel examined.

Authors' Reply

Mr. Vilella has presented, in the form of photomicrographs and reaction-rate curves, data that are in complete disagreement with the results we have obtained, and has reaffirmed his belief in the S-curve interpretation of the martensite process. Although Mr. Vilella states that his data are "clear-cut and accurate," he has neglected to include in his discussion any information regarding experimental procedure and it is, therefore, practically impossible



Figs. 1 to 4—Same Steel as that Studied by J. R. Vilella. Fig. 1—204 Degrees Cent., 1 Second. Fig. 2—204 Degrees Cent., 3 Minutes. Fig. 3—149 Degrees Cent., 1 Second. Fig. 4—149 Degrees Cent., 5 Minutes.

for a reader to estimate the value of his results. We have described in detail our method of study and had hoped that Mr. Vilella would include in his discussion a similar description of the method of microscopic study used for the low temperature range at the U. S. Steel Corporation Research Laboratory.

For some obscure reason, Mr. Vilella attributes the discrepancy in results to differences in impurity content of the alloys. Inasmuch as a substantial part of our paper deals with a study of a commercial drill rod steel, we can see no basis for Mr. Vilella's impression that our conclusions apply only to high

purity iron-carbon alloys. In order to make certain that the disparity in results as to the source could not be due to variations in the behavior of different steels, we asked Mr. Vilella to send us a sample of the steel used in the evaluation of reaction-rate curves presented in his discussion. To this request, Mr. Vilella kindly acceded, and in Figs. 1 to 4 are represented pertinent results from our brief study of this steel. Experimental details were identical to those described on pages 543 to 551 with the following exceptions: (1) The steel was received in the form of $\frac{1}{2}$ -inch square rod; (2) All specimens were of the same size, 0.035-inch square in cross section and $\frac{1}{4}$ -inch long, cut from the rod after $\frac{1}{8}$ -inch had been milled from each face of the original rod; (3) Only two quenching-bath temperatures were studied, namely 149 degrees Cent. and 204 degrees Cent. (corresponding to the two reaction-rate curves presented by Mr. Vilella); (4) Photomicrographs of the 204 degrees Cent. specimens were taken at $\times 150$ instead of $\times 500$ in order to record a larger and thus more representative area on the specimen.

The results obtained were those predictable from our interpretation of the martensite transformation outlined in the conclusion to the paper. For 204 degrees Cent., the 3-minute specimen contained little more decomposition product than did the 1-second specimen (see Figs. 1 and 2). For 149 degrees Cent., an increase of time from 1 second to 5 minutes resulted in no appreciable increase in the amount of martensite (see Figs. 3 and 4).

We are thus faced with the fact that the discrepancy in results from Mr. Vilella's study and from our own can be due only to differences in the methods of study. Though we by no means claim that our method is the only correct one, we do feel that each step in our experimental procedure is logical and that a reasonable amount of care has been exercised in evaluating experimental variables. A welcome and valuable contribution from Mr. Vilella would have been a detailed comparison of the two methods of study, apparently capable of producing from the same steel such widely differing results.

We are pleased to note that both Mr. Cohen and Mr. Scott are in agreement with our contention that the S-curve is in error for temperatures within the martensite range and should, therefore, be modified if it is to become the generally valuable diagram that it should be. The schematically revised diagram presented by Mr. Cohen may be of assistance in visualizing qualitatively the necessary modifications. We had considered the advisability of including a diagram similar to that presented by Mr. Cohen but decided to wait until austenite decomposition for the low-temperature range had been studied quantitatively. This quantitative study is now under way at Harvard, and we expect soon to be able to present reaction-rate curves for austenite decomposition within the martensite range, and a detailed modification of the S-curve.

TRANSFORMATION OF AUSTENITE ON QUENCHING HIGH PURITY IRON-CARBON ALLOYS

BY THOMAS G. DIGGES

Abstract

The tests described in this report were made primarily to show the influence of variation of distribution and solution of carbon in austenite on the mode and rate of transformation of austenite in high purity iron-carbon alloys. The transformation of a nonuniform grain of austenite often proceeds at various rates in different regions within the grain and its initial transformation to troostite is not always confined solely to its grain boundaries. These facts are illustrated by a representative case in which an individual grain of austenite transformed to cementite, ferrite, lamellar pearlite, nodular troostite and martensite. The results of microhardness tests made on the decomposition products of this same grain of austenite, the influence of grain size on the transformation rate of austenite, and the occurrence of cracks in quenched specimens of the hypereutectoid alloys are also discussed.

INTRODUCTION

THE cooling rate required to prevent austenite from transforming to troostite (fine pearlite) is influenced by the composition, grain size and chemical homogeneity of the austenite at the time of quenching. The important influence of the carbon content on the rate of transformation of austenite of both high purity iron-carbon alloys and plain carbon steels was discussed in a previous publication (1).¹ The previous study included the determination of the critical cooling rates of the alloys and steels quenched directly from temperatures establishing an average constant austenitic grain size and complete solution of the carbon, and in addition, the critical cooling rates of the plain carbon steels quenched from the usually recommended hardening temperature range. In carrying out the latter ex-

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. The author, Thomas G. Digges, is metallurgist, National Bureau of Standards, Washington, D. C. Manuscript received May 17, 1939.

periments, it was observed that some of the rapidly quenched specimens of the 0.20 per cent carbon steels contained areas of free ferrite. At the time of quenching these specimens, the carbon was nonuniformly distributed in the austenite, and regardless of the high cooling rates, the specimens were not completely hardened.

The present study was made to show the influence of distribution and solution of the carbon in austenite on the mode and rate of transformation of austenite on quenching high purity iron-carbon alloys. Determinations were also made to show the influence of grain size of austenite on its rate of transformation in these alloys. The qualitative results of the latter experiments will be supplemented by quantitative data in a subsequent report.

PREPARATION OF ALLOYS

The high purity iron-carbon alloys used in the present investigation were the same as those described in detail in a previous report (1). Essentially, the procedure for preparing these alloys consisted in carburizing hot- and cold-worked specimens of high purity iron in a mixture of hydrogen and benzene vapor, and subsequently homogenizing by heating in vacuo at 1700 degrees Fahr. (925 degrees Cent.). The method used in preparing the alloys eliminated the quality factor and variables other than carbon in the composition.

Results of spectrochemical and chemical analyses, and determinations of gas content by the vacuum fusion method, made on specimens of the iron-carbon alloys after treatment for homogeneity, showed that sulphur, nickel, cobalt, and oxygen were the major impurities contained in the alloys. These elements amounted to about 0.021 per cent, whereas the total percentage (by weight) of all impurities determined was about 0.030.

EXPERIMENTAL PROCEDURE

All specimens used in the present study were approximately 1/10 inch square by 0.040 inch thick. One wire of a 32-gage chromel-alumel thermocouple was spot-welded to the center of the flat face of the specimen and the other wire was spot-welded to the center of the opposite face. The specimens were heated rapidly in vacuo within a chromel coil, approximately 1 minute being required to reach the desired temperature, and held at that temperature for

15 minutes before quenching in hydrogen. Depending upon the purpose of the test, the specimens were either quenched directly from the high temperature or were cooled from this temperature to a temperature in the vicinity of A_1 before quenching. The assembly of the apparatus used for heating the specimens in vacuo and quenching directly in hydrogen is shown in Fig. 1.² The chamber (E) containing the heating coil and the specimen is evacuated before heating the specimen. On quenching, the stop-cock (D) to the vacuum pump is closed, the switch (B) opened and the stop-cock (F) turned to permit passage of hydrogen into the heating chamber. The hydrogen after passing around the specimen is exhausted into the air through the mercury seal (C). The desired cooling rates are obtained by regulating the flow of gas with reducing and needle valves attached to the hydrogen tank.

TRANSFORMATION OF AUSTENITE ON QUENCHING THE ALLOYS

Uniform Austenite. Mode of Transformation and Influence of Grain Size on Transformation Rate

The manner in which austenite³ transforms on quenching high purity iron-carbon alloys, and the influence of grain size on its rate of decomposition, are shown in Fig. 2. These structures were produced in alloys containing 0.73 per cent carbon (Fig. 2A) and 0.85 per cent carbon (Fig. 2B) by heating in vacuo to 1800 degrees Fahr. (980 degrees Cent.) and quenching directly in hydrogen at rates slower than the critical cooling rate. At the start of the quench, all carbon was in solution and uniformly distributed in austenite of mixed grain size.

It is evident that the initial transformation of each grain of austenite occurred at its boundary and that the interior of the austenitic grain was cooled unchanged to a temperature favorable for its transformation to martensite. That is, the austenite in the vicinity of the grain boundary was the least stable and, on quenching, transformed at a higher temperature than the more stable austenite in the interior of the grains.

The grain boundaries of the uniform austenite contained the

²This apparatus will be described in more detail in a subsequent report.

³The term "uniform austenite" as used in this report denotes complete solution and uniform distribution of carbon in gamma iron. Conversely, "nonuniform austenite" denotes incomplete solution or nonuniform distribution of carbon, or both, in gamma iron.

nuclei that were effective in the formation of nodular troostite during the quench. These nodules of troostite appear to have grown radially from the nuclei. The fact that a large proportion of these nodules is located symmetrically with respect to the grain boundaries of both the uniform austenite (Fig. 2) and nonuniform austenite

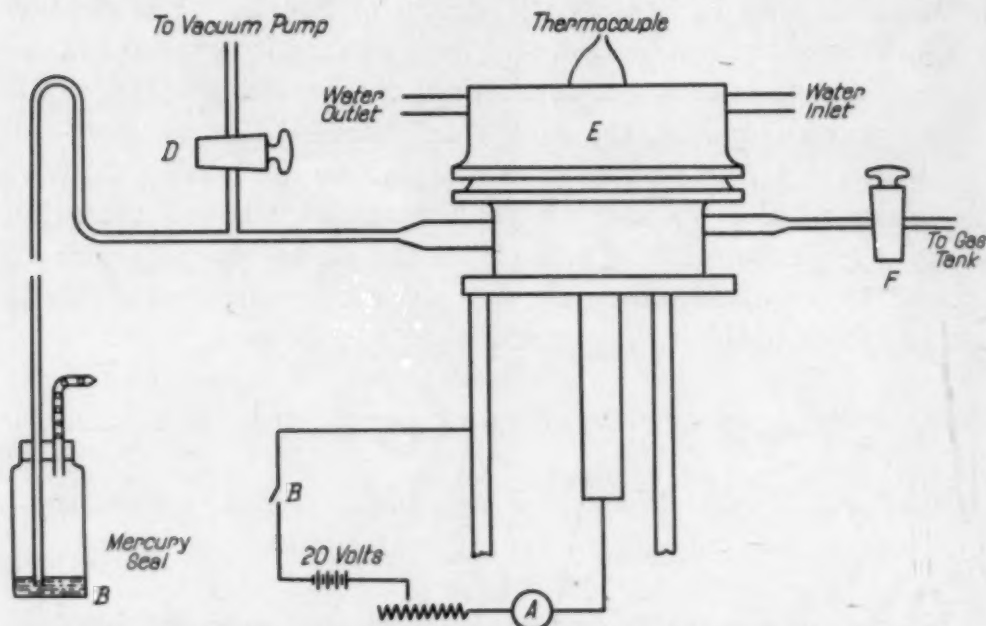


Fig. 1—Diagram of the Apparatus Used for Heating the Specimens in Vacuo and Quenching Directly in Hydrogen. A—Ammeter, B—Switch, C—Mercury Seal, D & F—Stopcocks, E—Heating Chamber.

(Fig. 5) indicates that a nucleus may be so located as to start transformation simultaneously in more than one grain of austenite. Possibly the majority of the nuclei, however, were located in such a manner as to restrict the activity of each nucleus to a particular austenitic grain.

Mehl (2) has discussed the decomposition of austenite of eutectoid composition. He expresses the opinion that the nuclei grow to approximately half spheroids, by a radial fan-like extension, until adjacent nodules come into contact whence they continue to grow by radial and columnar extension towards the center of the grain. He further states that nodules of pearlite (troostite) extend only into the grains in which the nucleus forms, and do not cross the austenitic grain boundary during growth. He considers that the process of the formation of martensite is definitely not one of nucleation and growth.

The relative stability of small and large grains of austenite is shown by comparing the decomposition products of these grains.

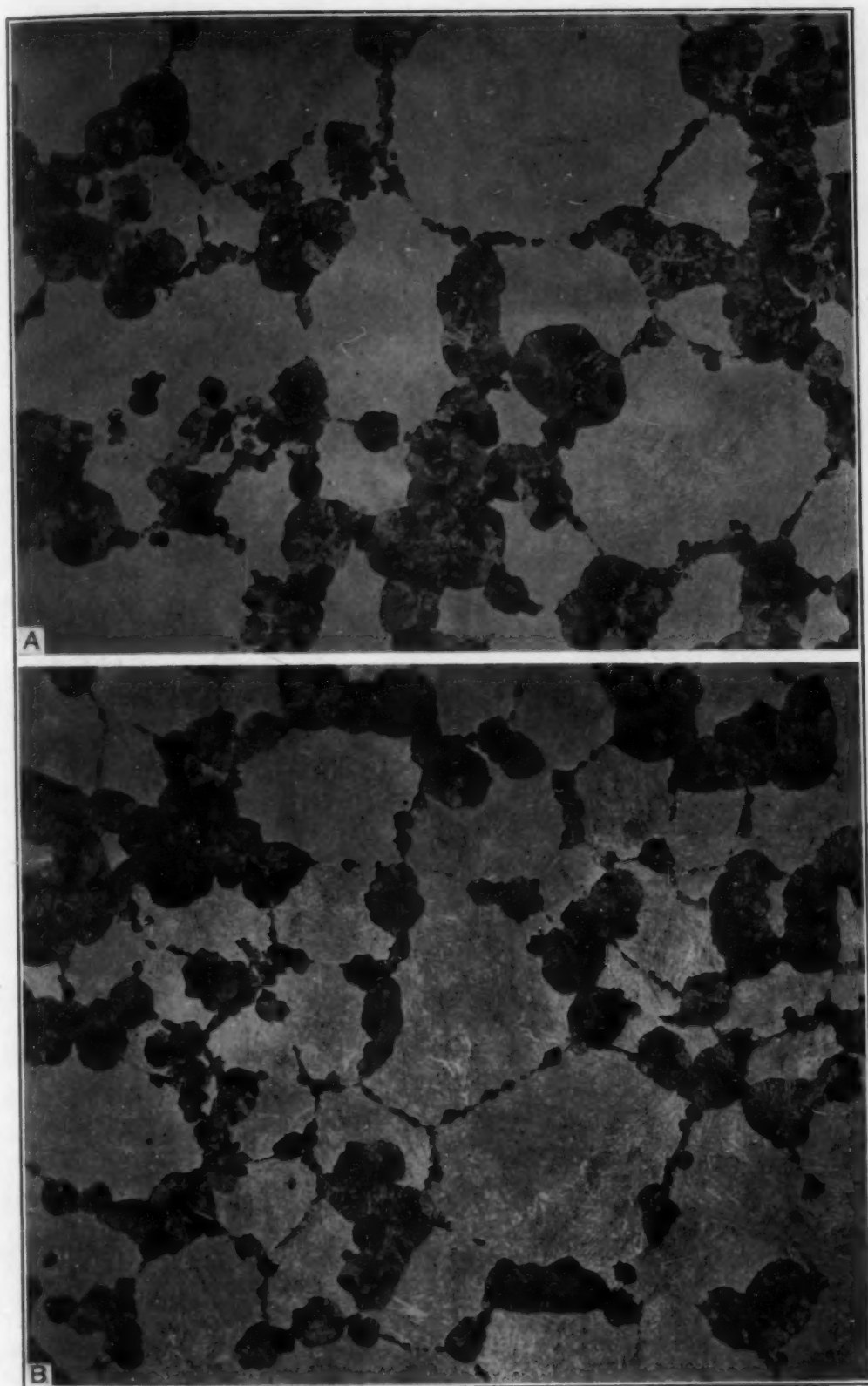


Fig. 2—Structure of Iron-Carbon Alloys Quenched in Hydrogen from 1800 Degrees Fahr. Fig. 2A—0.73 Per Cent Carbon. Fig. 2B—0.85 Per Cent Carbon. $\times 100$.

The smaller grains after quenching obviously contained larger proportions of troostite than did the similarly cooled large grains. With all other factors which affect the transformation rates remaining constant, the larger the grain size the more stable is the austenite or the deeper the hardening of the alloy. This is in agreement with the general concept of the relationship between austenitic grain size and hardenability. Quantitative data on the influence of austenitic grain size on the transformation rates of these alloys will be presented in a subsequent report.

The type of structure produced in the 0.23 per cent carbon alloy by quenching directly from 1800 degrees Fahr. (980 degrees Cent.) in hydrogen at a rate in excess of the critical cooling rate is shown in Fig. 3A. The austenite started to transform at about 890 degrees Fahr. (475 degrees Cent.) (Fig. 14), and its decomposition product consisted entirely of low carbon martensite. Fig. 3B shows the structure produced in the same alloy by quenching directly from 2100 degrees Fahr. (1150 degrees Cent.) in hydrogen at a rate slower than the critical cooling rate. The austenite transformed to proeutectoid ferrite at its grain boundaries and to martensite in the interior of the grains. This shows that when transformation of the 0.23 per cent carbon austenite occurred at temperatures above the range where martensite was formed, the transformation product was ferrite, not troostite, as was the case in higher carbon alloys.

Fig. 3C shows the structure produced in the 1.21 per cent carbon alloy by quenching directly from 1800 degrees Fahr. (980 degrees Cent.) in hydrogen at about the critical cooling rate. This specimen contained a large number of fine quenching cracks in the martensite. These cracks were located not only in the original austenitic grain boundaries but also in the interior of the grains. This condition was characteristic of the hypereutectoid alloys on quenching from high temperatures at rates which completely hardened the alloys. However, when sufficient external stress was applied to these martensitic specimens to cause rupture, the fractures were intergranular with respect to the parent austenite grains. This is illustrated in Fig. 4, representing specimens of the hypereutectoid alloys quenched directly from 1800 degrees Fahr. (980 degrees Cent.) in hydrogen, by the appearance of cracks after Rockwell hardness determinations were made. Fig. 4A shows the grain size (at 1800 degrees Fahr.) of the alloy containing 1.21 per cent carbon; this grain size is similar to the other alloys at this temperature.

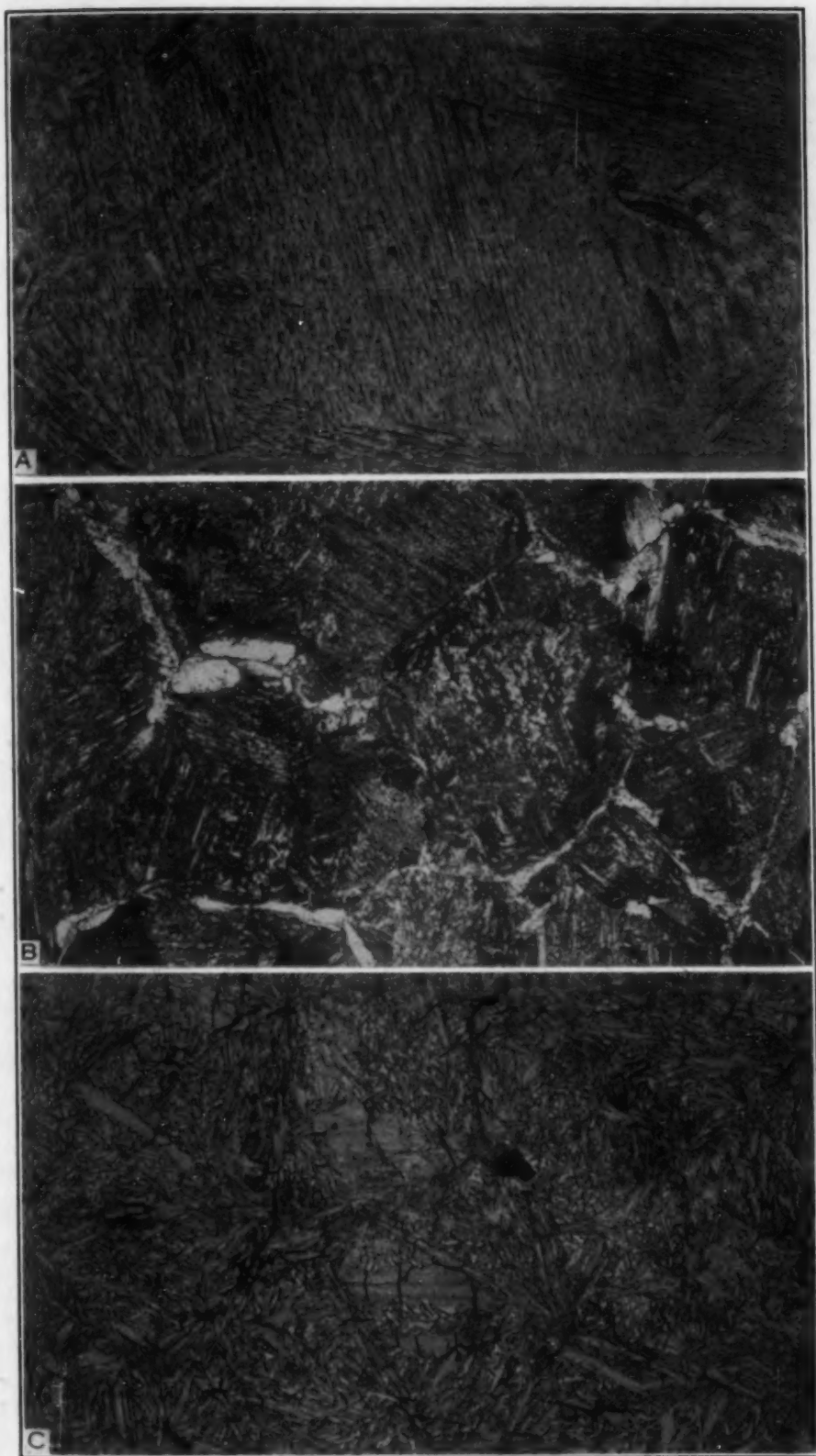


Fig. 3—Structure of Iron-Carbon Alloys Quenched in Hydrogen from 1800 Degrees Fahr. Fig. 3A—0.23 Per Cent Carbon, Rate in Excess of the Critical Cooling Rate, $\times 250$. Fig. 3B—0.23 Per Cent Carbon, Rate Slower Than Critical Cooling Rate, $\times 100$. Fig. 3C—1.21 Per Cent Carbon, Rate About the Critical Cooling Rate, $\times 250$.

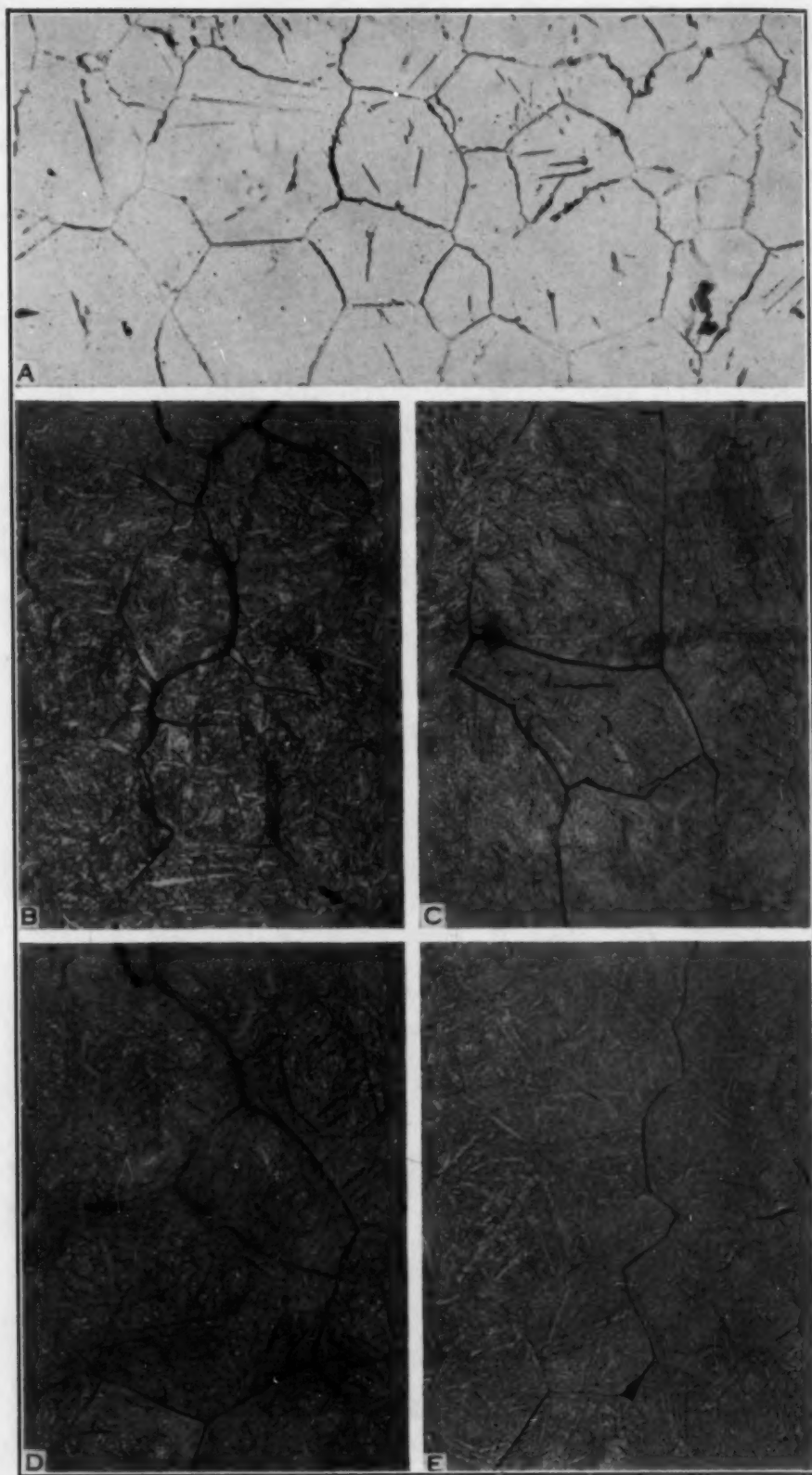


Fig. 4—Intergranular Cracks in Martensite of Iron-Carbon Alloys Quenched in Hydrogen from 1800 Degrees Fahr. Fig. 4A—1.21 Per Cent Carbon, $\times 100$; Fig. 4B—1.01 Per Cent Carbon, $\times 100$. Fig. 4C—1.01 Per Cent Carbon, $\times 250$. Fig. 4D—1.14 Per Cent Carbon, $\times 100$. Fig. 4E—1.21 Per Cent Carbon, $\times 100$.

Nonuniform Austenite. Mode of Transformation and Influence of Carbon Distribution and Grain Size on Transformation Rate

The effect of incomplete solution of carbon in austenite of varying grain size on the rate of transformation of the 1.14 per cent carbon alloy is shown in Fig. 5. The structure illustrated was obtained in a specimen that was heated in vacuo to 2000 degrees Fahr. (1095 degrees Cent.), then cooled to about 1330 degrees Fahr. (720 degrees Cent.) and held at the latter temperature for about 5 minutes before quenching in hydrogen.

The parent austenitic grains at 2000 degrees Fahr. (1095 degrees Cent.) were outlined not only with the cementite network precipitated by the treatment prior to quenching but also with the nodular troostite formed during the time of quenching. With the exception of a few of the grains of smaller size, each grain of austenite was identified by each of its decomposition products, cementite and troostite. On quenching, several of the smaller grains transformed completely to troostite, whereas the larger grains transformed predominantly to martensite. It should be noted, however, that at the time of quenching, approximately eutectoid proportion of carbon only was in solution and the manner in which the specimen was cooled did not necessarily insure the same distribution of the carbon in the small as in the large grains.

With this specimen, the mode of transformation of the austenite on quenching was similar to that already described for uniform austenite. The nuclei effective in the decomposition of the austenite to troostite were located in the boundaries of the grains. However, the initial transformation of nonuniform austenite to troostite does not necessarily occur solely in the grain boundaries.

The influence of variations of distribution and solution of carbon on the mode and rate of transformation of austenite of the high purity iron-carbon alloys is illustrated by the structures in Fig. 6 produced in specimens of the 1.21 per cent carbon alloy. The structure shown in Fig. 6A is that of a specimen quenched directly from 1800 degrees Fahr. (980 degrees Cent.) in hydrogen; the structure shown in Fig. 6B is that of a specimen heated to 1800 degrees Fahr. (980 degrees Cent.) and cooled to 1380 degrees Fahr. (750 degrees Cent.) (a temperature above A_1) and held at this temperature for 5 minutes before quenching in hydrogen; and the structure shown in Fig. 6C is that of a specimen heated to 1800 degrees Fahr. (980

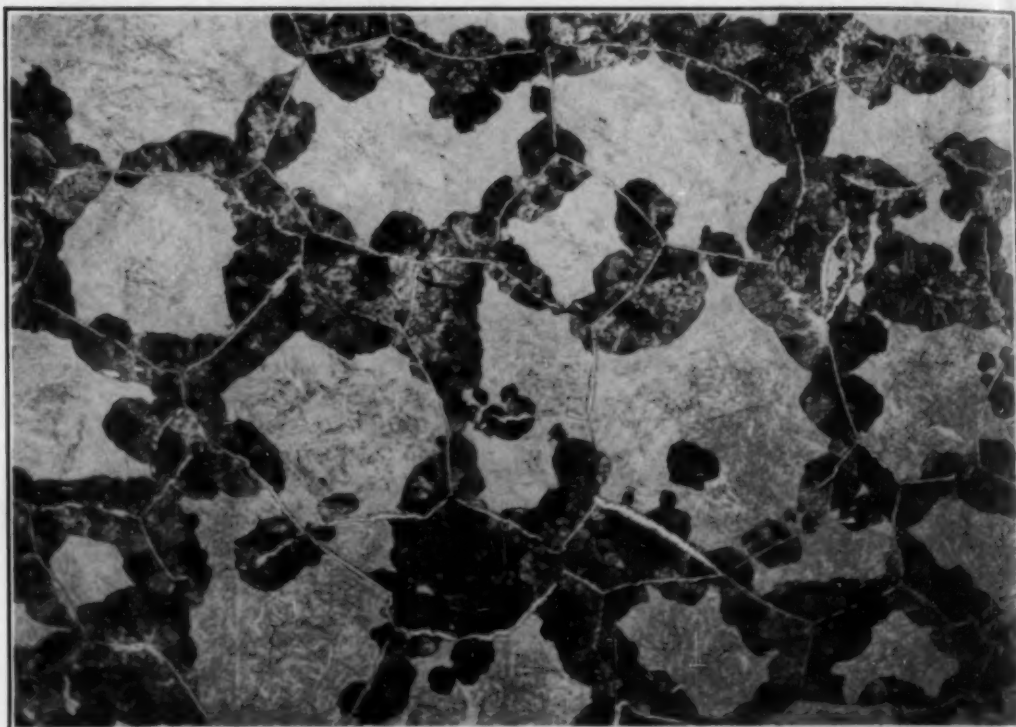


Fig. 5—Grain Size at 2000 Degrees Fahr. and Structure of Quenched Iron-Carbon Alloy Containing 1.14 Per Cent Carbon.

Specimen Was Cooled from 2000 to 1330 Degrees Fahr. and Held at the Latter Temperature for 5 Minutes Before Quenching in Hydrogen. The Grain Size at 2000 Degrees Fahr. is Shown by the Cementite Precipitated at the Austenitic Grain Boundaries by the Treatment Prior to Quenching and Also by the Troostite Formed at the Grain Boundaries During the Quench. Except for Several of the Small Austenitic Grains that Transformed Entirely to Troostite, the Same Austenitic Grains Were Revealed by Each of its Decomposition Products Consisting of Cementite and Troostite. Etched with 1 Per Cent Nitric Acid in Alcohol. $\times 100$.

degrees Cent.) and cooled to about 1330-1310 degrees Fahr. (within the temperature range of Ar_1) and held in this temperature range for about 8 minutes before quenching in hydrogen. All these specimens, therefore, had the same average grain size of the parent austenite and the same total carbon content. Each specimen was cooled through the temperature range 1110 to 930 degrees Fahr. (600 to 500 degrees Cent.) (the upper range in which the austenite is least stable) at an average rate of approximately 520 degrees Fahr. (290 degrees Cent.) per second. The specimen quenched directly from 1800 degrees Fahr. (980 degrees Cent.) evidently cooled practically unchanged to the low temperature favorable for the decomposition of austenite to martensite. The average hardness value of this specimen was Rockwell C-62. The specimen quenched from 1380 degrees Fahr. (750 degrees Cent.) had a structure consisting of proeutectoid cementite and a small amount of nodular troostite in a matrix of martensite. The average hardness of this specimen

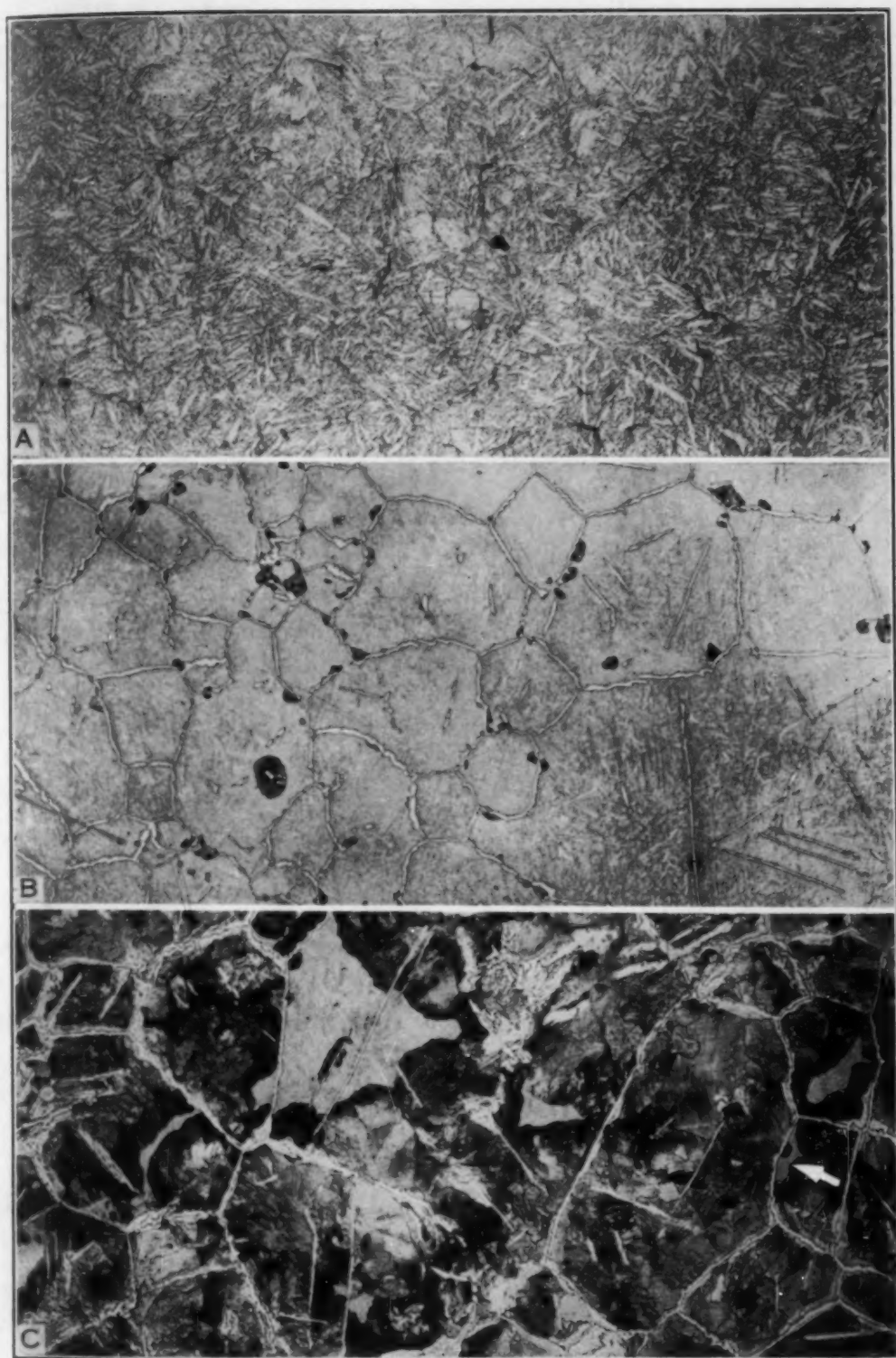
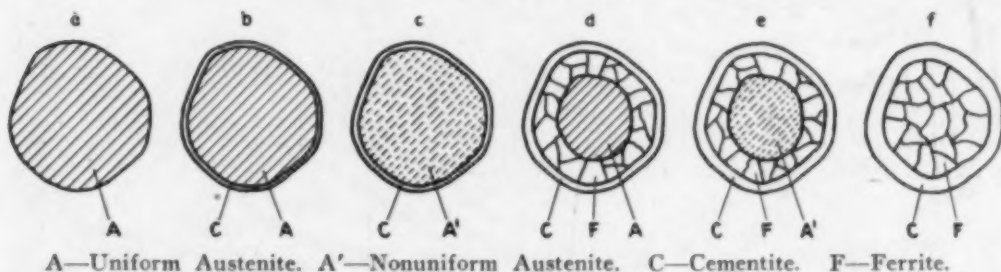


Fig. 6—Structure of Iron-Carbon Alloys Quenched in Hydrogen from 1800 Degrees Fahr. Specimens Were 0.10 Inch Square by 0.040 Inch Thick, Cooled from 1110 to 930 Degrees Fahr. at 520 Degrees Fahr. per Second. $\times 100$.

was Rockwell C-65. The specimen quenched from 1330-1310 degrees Fahr. (720-710 degrees Cent.) transformed over a wide range of temperature to ferrite, carbide and martensite, resulting in an average hardness value of Rockwell C-27. Although the specimen which was cooled to 1380 degrees Fahr. (750 degrees Cent.) before quenching contained more troostite than the one quenched directly from 1800 degrees Fahr. (980 degrees Cent.), it is noteworthy that the former specimen had the higher average hardness value (Rockwell C-65 and 62, respectively). The lower hardness value of the latter specimen is believed to be due to the retention of some austenite after the quench.

At the time of quenching the specimens, variations existed in the distribution and amounts of dissolved and undissolved carbon, as is shown schematically in Fig. 7. All carbon was in solution and uniformly distributed in the specimen quenched directly from 1800



A—Uniform Austenite. A'—Nonuniform Austenite. C—Cementite. F—Ferrite.
Fig. 7—Sketch Illustrating the Solution and Distribution of Carbon in the Hyper-eutectoid Alloys at the Time of Quenching.

degrees Fahr. (980 degrees Cent.) (Fig. 6). On cooling the other specimens from 1800 degrees Fahr. through the A_{cm} temperature range, proeutectoid cementite precipitated at the boundaries of the parent austenitic grains. At the time of quenching the specimen from 1380 degrees Fahr. (750 degrees Cent.), approximately eutectoid proportion of the carbon was in solution and the remaining undissolved carbon existed principally as grain boundary network. At the time of quenching the specimen from 1330-1310 degrees Fahr. (720-710 degrees Cent.), partial transformation of the austenite evidently had already taken place, and there existed an aggregate of austenite, carbide and ferrite.

The stability of the parent austenite decreases (hardenability decreases) progressively as the conditions represented in Fig. 7 change consecutively from *a* to *f*. Obviously, complete hardening of the alloy cannot be obtained by quenching the parent austenite when conditions represented by *d*, *e*, and *f* exist.

Quenching the specimens, referred to in Fig. 6, from different temperatures influenced the stability of the austenite, on cooling through the temperature range 1110 to 930 degrees Fahr., only to the extent of varying the amount and distribution of the carbon in solution at the time of quenching. This fact was confirmed by determinations of the critical cooling rate made on specimens of the 0.80 and 0.85 per cent carbon alloys. Specimens from each of these alloys were quenched directly from 1800 degrees Fahr. (980 degrees Cent.), while similar specimens were cooled from 1800 to 1400 degrees Fahr. (980 to 760 degrees Cent.) before quenching in hydrogen. Each specimen had the same average austenitic grain size and all carbon was in solution at the quenching temperatures of 1800 and 1400 degrees Fahr. The critical cooling rate for each alloy quenched directly from 1800 degrees Fahr. was the same as that for the corresponding alloy heated to 1800 degrees Fahr. and cooled to 1400 degrees Fahr. before quenching.

As shown in Fig. 6B, the initial transformation of nonuniform austenite to nodular troostite is not confined to the grain boundary region, but often occurs in the interior of the parent grains.

The mode of transformation of austenite in high purity iron-carbon alloys also may be followed readily by allowing partial transformation of the austenite before quenching. Transformation products obtained in portions of two adjacent grains of nonuniform austenite (grains indicated by arrow at right of Fig. 6C) of such a specimen of the alloy containing 1.21 per cent carbon are reproduced in Fig. 8A. At the time of quenching, these two grains were separated from each other by cementite. Along this common boundary, the smaller austenite grain transformed to martensite and nodular troostite, whereas the larger grain transformed to lamellar pearlite (see Fig. 6C for relative size of these two grains). Martensite is the transformation product of the relatively stable austenite, troostite is that of the less stable austenite, and lamellar pearlite is that of the least stable austenite.

Similar distribution of the transformation products of austenite was observed not only in other parts of this same specimen but also in another specimen of the 1.21 per cent carbon alloy similarly treated. Indeed, an area of lamellar pearlite entirely surrounded by troostite (fine pearlite) was found in the interior of one of the parent austenitic grains. It is surprising to find martensite at the grain boundary and troostite in the interior of a relatively small

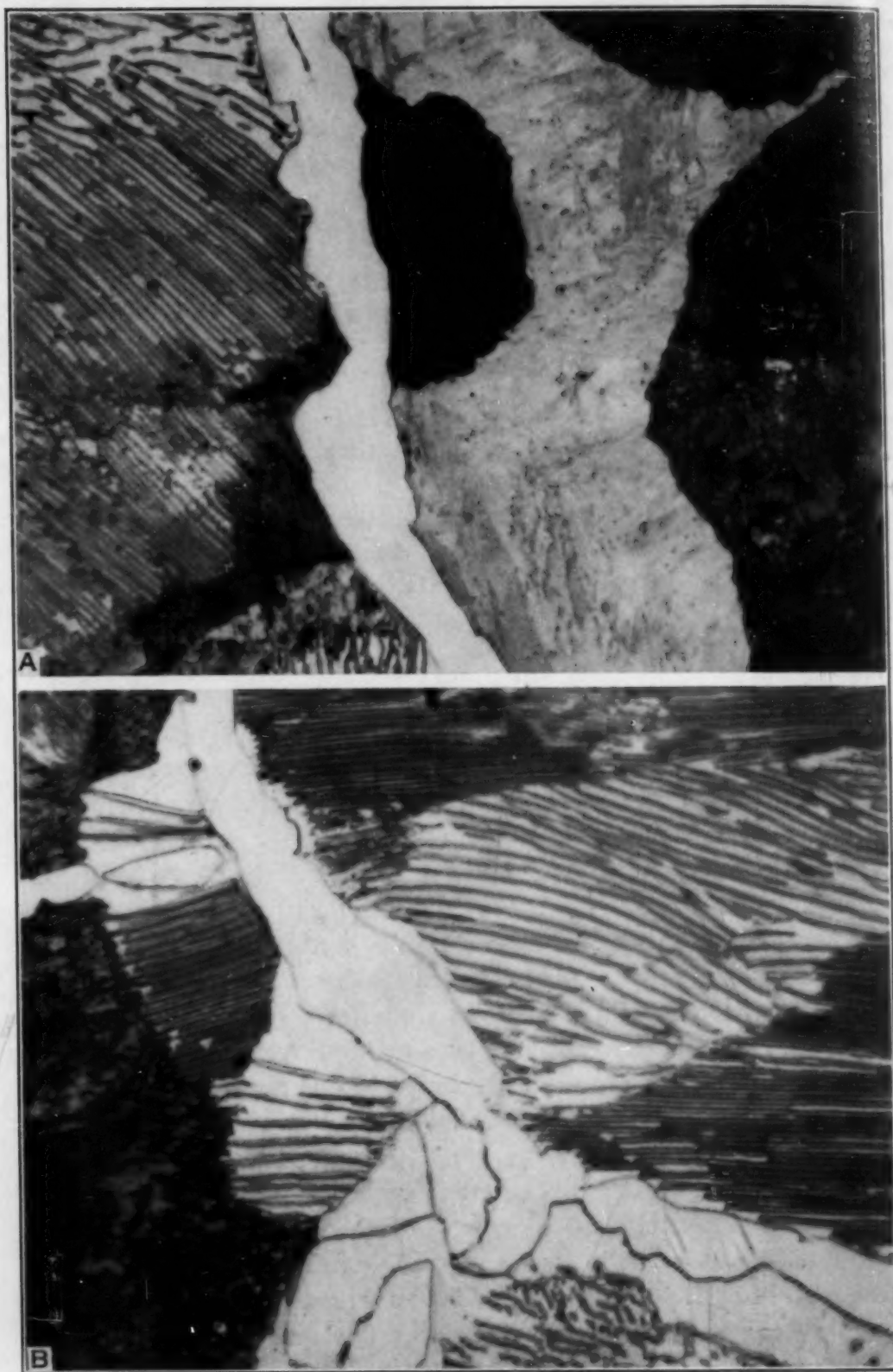


Fig. 8—Structure of Quenched Iron-Carbon Alloy Containing 1.21 Per Cent Carbon. Etched with 1 Per Cent Nitric Acid in Alcohol. $\times 2000$.

grain of the parent austenite and lamellar pearlite in the adjacent large grain, the pearlite being separated from the martensite only by the common grain boundary constituent of cementite. The actual cause of such a mode of transformation of austenite cannot be definitely stated at this time. It is recognized that the conditions were favorable for carbon segregation prior to quenching, and this is believed to have been a contributing factor. These results show that the initial transformation of austenite does not always take place only at its grain boundary and that the grain size of the austenite is not always a true index to its stability.

The hypereutectoid alloys had a high degree of abnormality in structure as shown by coalesced masses of cementite and ferrite (3). This abnormality was observed not only in the annealed alloys but also in some of the quenched structures (Figs. 10 and 12). An interesting feature, therefore, was the absence of abnormality in the structure shown in Fig. 8A. That abnormality did exist in portions of this same specimen is shown by the structure of Fig. 8B.

A specimen of the alloy containing 1.14 per cent carbon was heated to 1700 degrees Fahr. (925 degrees Cent.), cooled to 1330-1310 degrees Fahr. (720-710 degrees Cent.), and held in the latter temperature range 5 minutes before quenching in hydrogen. Fig. 9A shows the grain size of the alloy at 1700 degrees Fahr. (925 degrees Cent.). With the boiling sodium picrate/etching reagent, the cementite in the grain boundaries was attacked and the martensite was darkened. Only the large grains contained the dark massive constituent (martensite), thus illustrating the relationship between grain size and hardenability. However, the large grain, shown in Fig. 9B, behaved as though grain size was not the only factor influencing its stability. The parent austenitic grain is outlined by a cementite envelope. Adjacent to the cementite is a layer of ferrite. Lamellar pearlite and varying degrees of fine pearlite extend from the ferrite to the large area of martensite (see also Fig. 10). Enclosed by the martensite is a light constituent that appears to be only mildly attacked with boiling sodium picrate. If the center of the original austenitic grain was the most stable part of the grain, then this constituent should be retained austenite. That it is nodular troostite and not retained austenite will be shown later. Another interesting feature shown in the decomposition products of this large grain of austenite is the small area of martensite formed a short distance from the grain boundary. This result is contrary

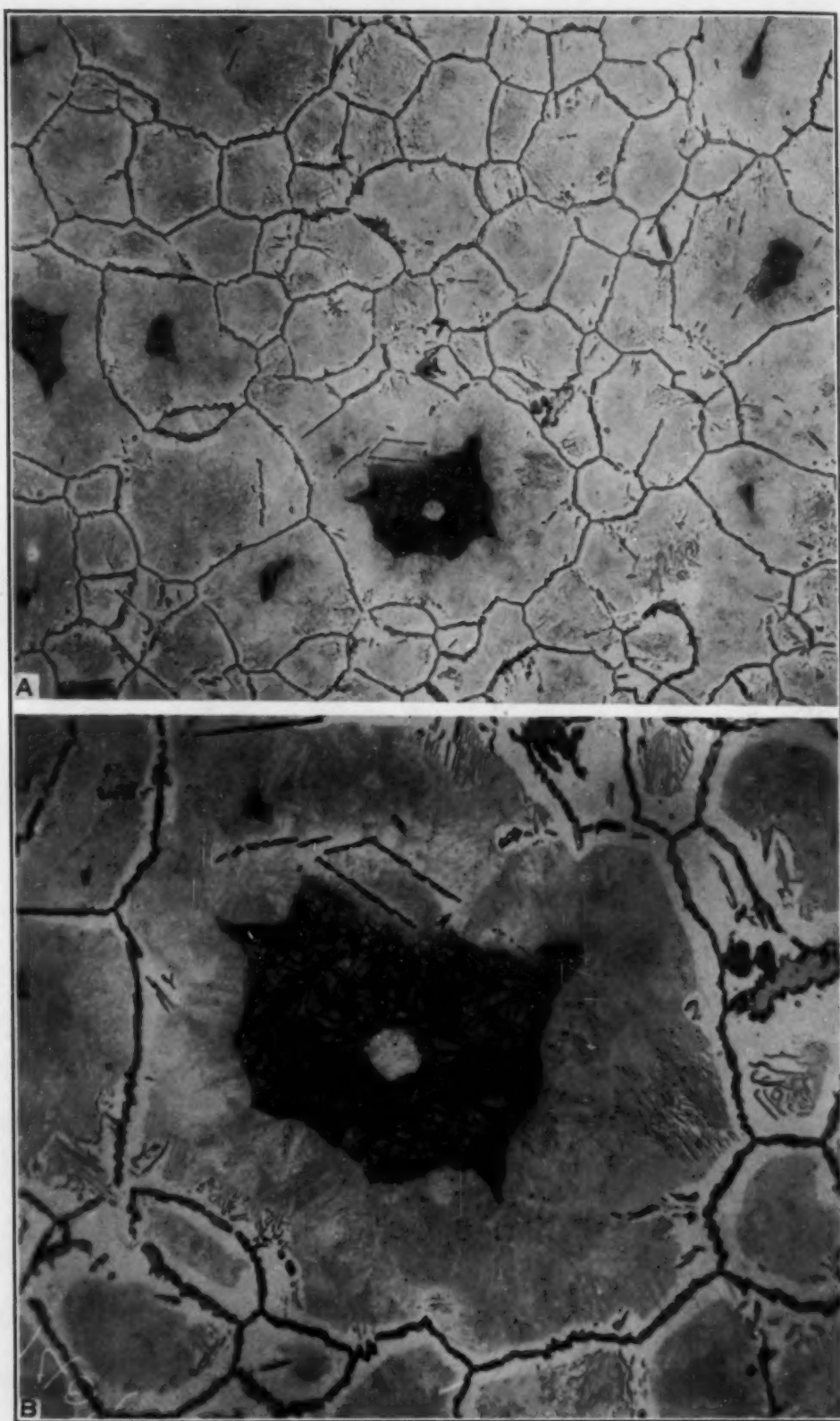


Fig. 9—Grain Size at 1700 Degrees Fahr. and Structure of Quenched Iron-Carbon Alloy Containing 1.14 Per Cent Carbon. Etched with Boiling Sodium Picrate. Fig. 9A, $\times 100$; Fig. 9B, $\times 250$.

to the generally accepted idea that the transformation of an austenitic grain occurs initially at its boundary and proceeds progressively toward its center.

To identify the constituent formed in the interior of the large area of martensite, the specimen was lightly repolished and subsequently etched in 1 per cent nital. The resulting structure is shown in Fig. 10. With the nitric acid etch, this particular constituent was darkened, thus showing that it was nodular troostite. This finding was further confirmed by the results of microhardness tests, Fig. 11.

The microhardness tests were made with an instrument using a diamond indenter of a special shape which was developed and described by Knoop, Peters, and Emerson (4). The microhardness number, H , was computed from the formula

$$H = \frac{\text{Load (Kg)}}{\text{Projected Area (mm}^2\text{)}}$$

Impressions designated 1 to 9 in Fig. 11 were made with a load of 100 g., while those shown near the left of the original austenitic grain were made with a load of 50 g. For the range in hardness shown, the microhardness and Brinell numbers agree sufficiently closely to permit consideration of the present values in terms of the more familiar Brinell numbers. The nodule of troostite, 25 to 30 microns in diameter (position 4), had a hardness number of 570 whereas the surrounding martensite (positions 3, 5 and 6) had a value of about 700. Determinations of hardness were also made on the coarse lamellar pearlite grain both transversely (position 8) and in the direction of the laminations (position 9). The results showed that the pearlite grain was appreciably softer in the direction of the laminations.

An interesting feature brought out by the microhardness tests was the wide variations in values obtained for the decomposition products of one grain of austenite although the cooling rate must have been uniform throughout the entire grain. The hardness values determined ranged from about 190 for the soft pearlite to about 720 for the fully hardened martensite. Determinations of hardness were not made in areas consisting entirely of ferrite or of cementite.

There remains no doubt as to the identification of the transformation products of this quenched grain of austenite. However, there is need for further discussion of the factor believed to have

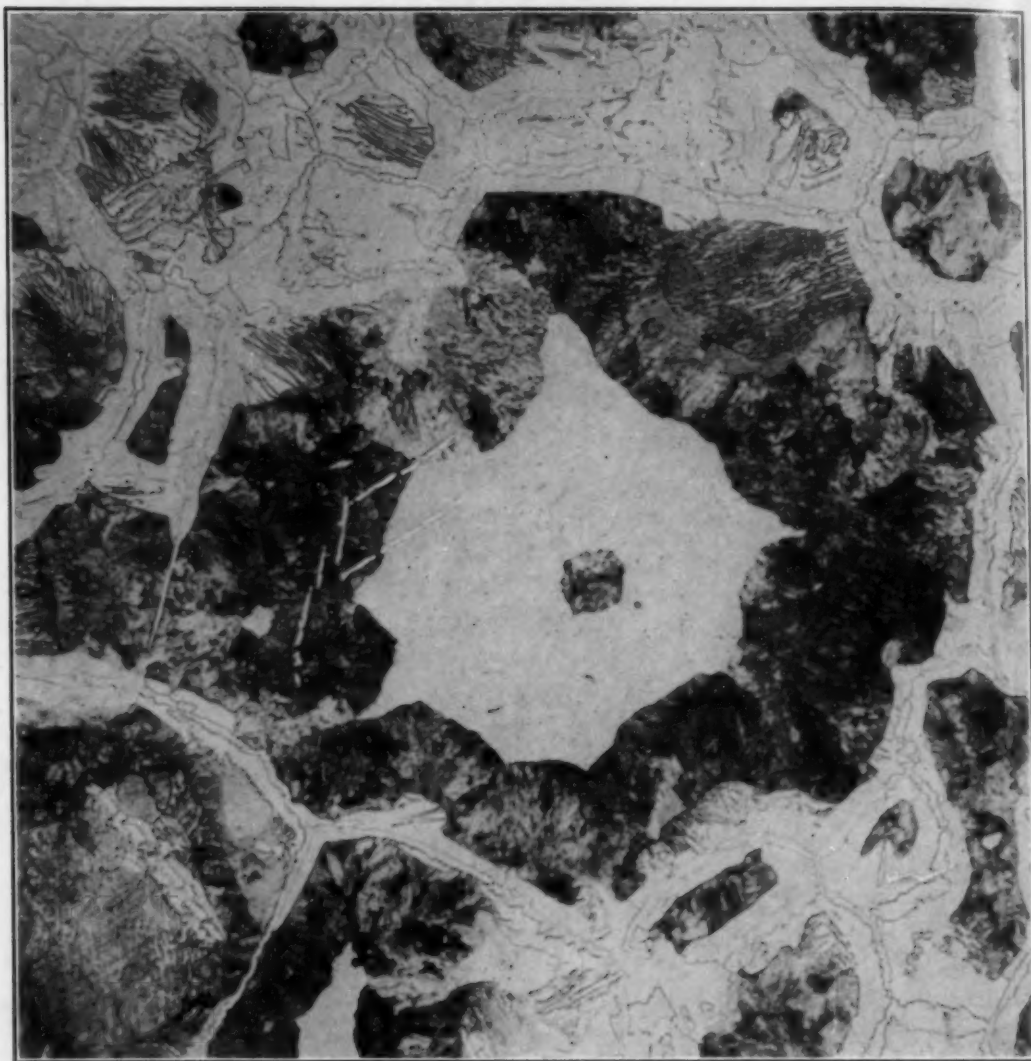


Fig. 10—Transformation of Austenite in Iron-Carbon Alloy Containing 1.14 Per Cent Carbon.

Decomposition Products of the Same Large Grain of Austenite as Shown in Fig. 9B. The Large Grain of Austenite Transformed to Cementite, Ferrite, Lamellar Pearlite, Troostite and Martensite. Stability Within This Grain of Nonuniform Austenite Did Not Increase Uniformly with the Distance from the Grain Boundary. At the Time of Quenching, Some of the Small Grains Consisted Essentially of Coalesced Cementite and Ferrite. Regardless of the Rate of Cooling, These Grains Cannot be Completely Hardened. Etched with 1 Per Cent Nitric Acid in Alcohol. $\times 250$.

been primarily responsible for the wide variations in its reaction rates. These variations obviously cannot be attributed to either differences in rates of cooling from the boundary to the center of the grain or to grain size. Furthermore, spectrochemical analysis made on an adjacent specimen of this 1.14 per cent carbon alloy showed no indication of the presence of aluminum. Evidently alumina inclusions, to act as nuclei for its transformation, were not present in the austenite. However, the manner in which the austenite

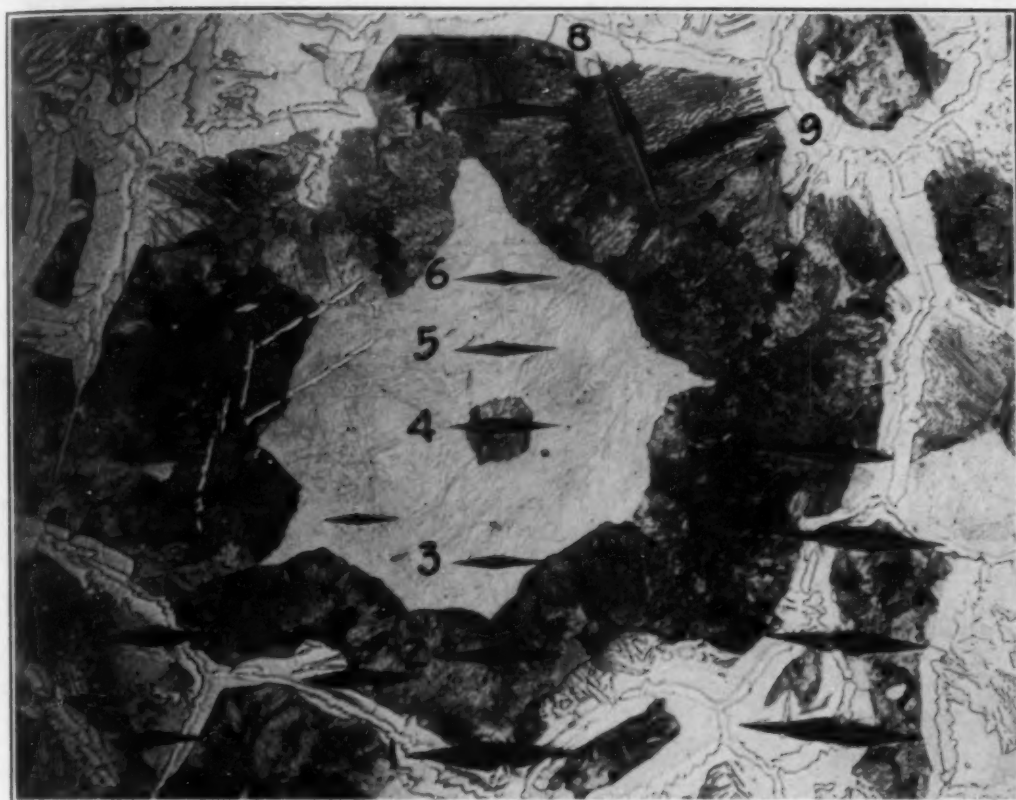


Fig. 11—Microhardness of Decomposition Products of a Grain of Austenite of Iron-Carbon Alloy Containing 1.14 Per Cent Carbon.
Transformation Products of the Same Large Grain of Austenite as Shown in Figs. 9B and 10. Etched with 1 Per Cent Nitric Acid in Alcohol. $\times 250$.

Position	Microhardness ⁽¹⁾ number	Position	Microhardness ⁽¹⁾ number
1	260	6	725
2	475	7	425
3	690	8	300
4	570	9	190
5	720

⁽¹⁾For the range of hardness shown, the microhardness and Brinell numbers agree sufficiently closely to permit consideration of the present values in terms of the more familiar Brinell numbers.

was cooled was very favorable for segregation of carbide in the grain at the time of quenching. It seems probable, therefore, that carbon gradients were largely responsible for the wide range in decomposition products of this uniformly quenched grain of austenite.

In a previous report (1) it was shown that with constant austenitic grain size and with all carbon in solution in the austenite, the critical cooling rate decreases (stability of austenite increases) continuously with increase in carbon for both these high purity iron-carbon alloys and plain carbon steels.

Under certain conditions, the particles of free carbides at the time of quenching tend to accelerate the decomposition of the austen-



Fig. 12—Structure of Quenched Iron-Carbon Alloy Containing 1.21 Per Cent Carbon. Specimen Was Cooled from 1700 Degrees Fahr. to About 1330 Degrees Fahr. and Held at the Latter Temperature for 5 Minutes Before Quenching in Hydrogen. A Nodule of Troostite Was Formed Around the Free Carbide in the Large Austenitic Grain Shown in the Top Portion of the Micrograph, Whereas in the Large Adjacent Grain Directly Below There Were Several Free Carbide Particles Without the Presence of Troostite in the Martensitic Matrix. At the Time of Quenching, Some of the Small Grains Consisted Essentially of Coalesced Cementite and Ferrite. Etched with 1 Per Cent Nitric Acid in Alcohol. $\times 250$.

ite by serving as centers for its transformation during cooling. Under other conditions, the free carbides apparently have but little influence on the transformation rate of quenched austenite. These facts are illustrated in Fig. 12 by the structure produced in a specimen of the 1.21 per cent carbon alloy heated to 1700 degrees Fahr. (925 degrees Cent.) and then cooled to about 1330 degrees Fahr. (720 degrees Cent.) before quenching in hydrogen. A nodule of troostite was formed around the free carbide in the large austenitic grain, shown

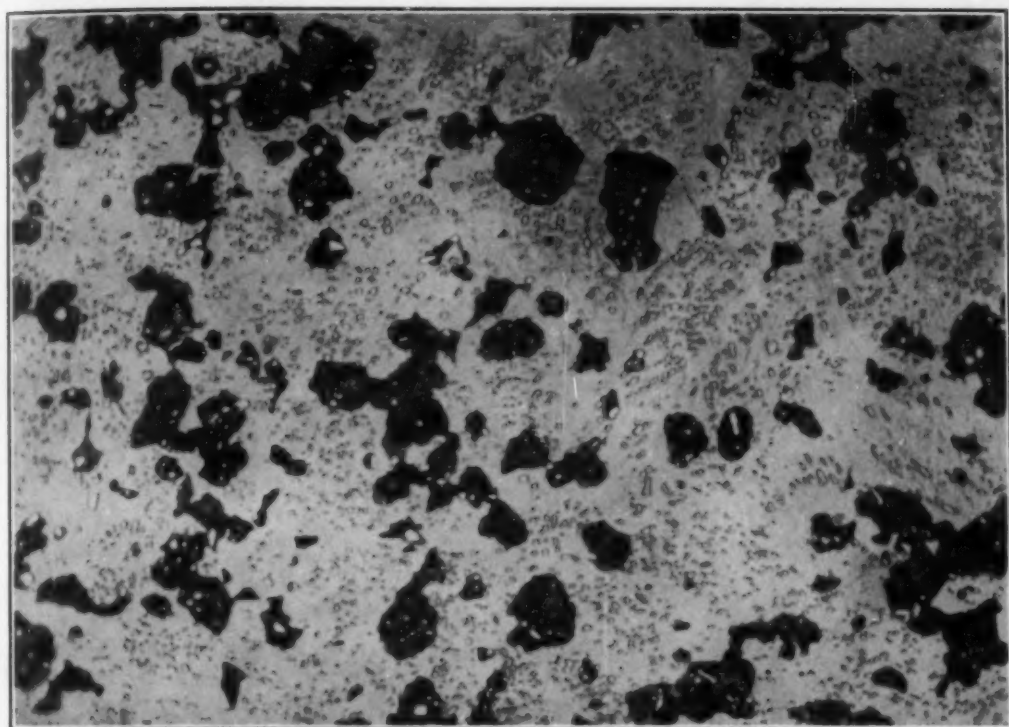


Fig. 13—Structure of Quenched Iron-Carbon Alloy Containing 1.01 Per Cent Carbon. Specimen Was Quenched Directly from 1425 Degrees Fahr. in Hydrogen. The Transformation Rate of the Austenite on Quenching Was Influenced by the Presence of Free Carbides. Etched With 1 Per Cent Nitric Acid in Alcohol. $\times 750$.

in the top portion of the micrograph, whereas in the large adjacent grain directly below there were several free carbide particles, without the presence of troostite, in the martensitic matrix. Furthermore, the carbide needles or plates that formed along certain planes within the original austenitic grain apparently did not influence the rate of decomposition of the austenite.

That the presence of free carbide in the austenite at the time of quenching influences its transformation rate is also shown in Fig. 13 by the structure obtained in a specimen of the 1.01 per cent carbon alloy heated to 1425 degrees Fahr. (775 degrees Cent.) and quenched directly in hydrogen. It is apparent that some of the undissolved carbide particles or free carbides acted as nuclei for transformation during cooling, and thus increased materially the reaction rates of the austenite. In Fig. 13 the martensitic matrix also contains numerous free carbide particles which did not act as transformation centers for austenite. This suggests that the reaction rate of the austenite increases with increase in the number of free carbides until a value is finally reached beyond which further increase in number has no effect on its transformation rate or hardenability. The transforma-

tion rate of austenite is believed to be influenced not only by the number and size but also by the distribution of the free carbides.

Morris and McQuaid (5) expressed the opinion that the actual differences in transformation rates or hardenability between fine- and coarse-grained austenite may well consist in low carbon or carbonless boundaries. That is, the relation between the physical size of the austenitic grain at the time of quenching and hardenability was thought to be less important than low carbon austenitic areas present because of nonuniform carbon solution and diffusion.

It has been shown that nonuniform distribution of carbon and the presence of free carbides in the austenite at the time of quenching materially influences the transformation rates of high purity iron-carbon alloys. These factors may have an effect equal to or even greater than that of grain size in controlling the reaction rates of the quenched austenite.

Influence of Carbon Content of Austenite on Temperature of the Ar'' Transformation

Specimens of the alloys ranging in carbon from 0.23 to 1.21 per cent were heated in vacuo to 1800 degrees Fahr. (980 degrees Cent.) and quenched directly in hydrogen. Each of the alloys had an average austenitic grain size of 1.5 to 3 grains per square inch at 100 diameters (A.S.T.M. grain number 2) at 1800 degrees Fahr. and all carbon was in solution at the time of quenching. The specimens were quenched at various rates, which were sufficiently fast to cool the austenite unchanged to the temperature range favorable for its transformation to martensite (Ar''). A photographic time-temperature cooling curve was obtained during the quench by means of a string-galvanometer apparatus.

The photographic records show clearly the arrest in the cooling curves at the Ar'' transformation of the hypoeutectoid and eutectoid alloys. With the type of cooling curves obtained in the present experiments no arrest was observed at the Ar'' transformation for the relatively stable austenite of the hypereutectoid alloys.

As shown in Fig. 14, the temperature at the start of the Ar'' transformation progressively decreased with increase in carbon content of the alloys from about 900 degrees Fahr. (480 degrees Cent.) for the 0.23 per cent carbon to about 500 degrees Fahr. (260 degrees Cent.) for the 0.80 per cent carbon alloy.

Esser, Ellender, and Spenle (6) found that the temperature

range of the Ar'' transformation was influenced both by the rate of cooling and by the carbon content of high purity iron-carbon alloys. Their results showed that the temperature of the Ar'' was lowered, with increase in carbon content, from 750 degrees Fahr. for the 0.20 per cent carbon to 170 degrees Fahr. for the 1.8 per cent carbon alloy. For plain carbon steels, French and Klopsch (7) showed that the temperature of Ar'' was lowered, with increase in carbon con-

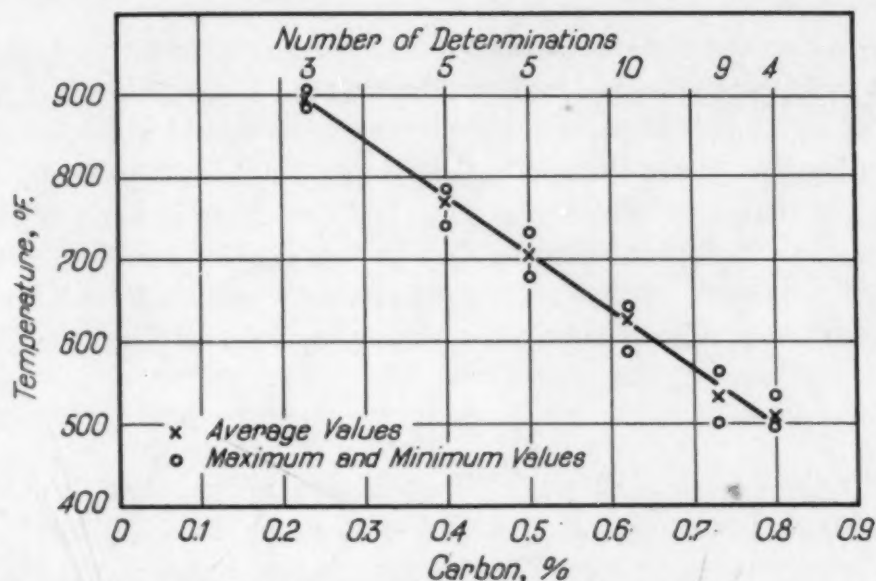


Fig. 14—Effect of Carbon Content on the Temperature of the Ar'' Transformation of Quenched Iron-Carbon Alloys.

Specimens 0.10 Inch Square by 0.040 Inch Thick Were Quenched in Hydrogen at Various Rates Directly from 1800 Degrees Fahr. At 1800 Degrees Fahr., All the Alloys Had the Same Average Grain Size of $1\frac{1}{2}$ to 3 Grains Per Square Inch at 100 X, and All Carbon Was in Solution.

tent, from about 1080 degrees Fahr. (580 degrees Cent.) for the 0.20 per cent carbon to about 710 degrees Fahr. (375 degrees Cent.) for the 0.60-0.70 per cent carbon and remained at about 710 degrees Fahr. for steels containing about 0.70 to 1.25 per cent carbon.

Davenport, Bain (8) and co-workers (9), in their studies of the process and results of austenite transformation at constant temperature, found that the general form of the transformation temperature-time curve was the same for all steels investigated, but that there were significant differences in the position of the S-curve on the time scale and minor differences on the temperature scale. The hardenability of a steel thus depends on the stability of its austenite on cooling through the temperature range of about 1110 to 930 degrees Fahr. (600 to 500 degrees Cent.) and the position of this range is not appreciably altered by variation in such factors as composition and

grain size of the austenite. In the present investigation, the A_r'' transformation commenced at temperatures appreciably higher than in the interrupted quench experiments of Davenport and Bain. Indeed, the A_r'' for the high purity iron-carbon alloy containing 0.23 per cent carbon began at about 900 degrees Fahr., a temperature only slightly below the upper range, as shown in the S-curve, where the austenite is unstable and in which its decomposition products are relatively soft.

Data on the transformation of austenite at constant temperature levels in high purity iron-carbon alloys would be of value in removing the existing confusion. This information would also aid in the determination of the critical cooling rates by definitely establishing the upper range in which the quenched austenite is most likely to decompose. However, accurate data on this subject would be difficult to obtain because of the relatively high rate of transformation of austenite in the high purity iron-carbon alloys, especially those of the lower carbon contents.

SUMMARY

Specimens of high purity iron-carbon alloys were heated in vacuo to various temperatures and either quenched directly from these temperatures or cooled to the temperature range in the vicinity of A_1 before quenching in hydrogen. The preparation and analyses of the alloys were described in a previous report (1).

Examination of the microstructures of the quenched specimens showed (a) the mode of transformation of austenite of varying grain size with both uniform and nonuniform distribution of the carbon, and (b) the effects of these factors on the rate of transformation of the austenite.

With all carbon in solution and uniformly distributed in austenite of mixed grain size, the initial transformation occurred at the grain boundaries. This austenite was least stable in the vicinity of its grain boundaries where transformation occurred at a higher temperature than in the interior of the grains.

With incomplete solution and nonuniform distribution of carbon in austenite of relatively large grain size, the transformation often proceeded at various rates within the grains. This was illustrated in an alloy containing 1.14 per cent carbon, by the transformation products of one large grain of austenite. This grain of austenite transformed to cementite, ferrite, coarse lamellar pearlite, varying

degrees of fine pearlite (troostite) and martensite. Microhardness tests made on this decomposed grain of austenite showed how the hardness varied with the transformation products. The microhardness values determined ranged from about 190 for the soft pearlite to about 720 for the fully hardened martensite.

Under certain conditions, free carbide particles in the austenite at the time of quenching act as nuclei for transformation and thus increase materially the reaction rate of the austenite. The reaction rate of the austenite apparently increases with increase in the number of free carbide particles per unit volume up to a point beyond which further increase in number has no effect on its transformation rate or hardenability.

The transformation rate of the alloys was influenced by the grain size of both uniform and nonuniform austenite with respect to carbon. With austenite of constant carbon content, uniformly distributed, the larger the grain size the more stable is the austenite (the slower is its transformation rate on quenching). With nonuniform distribution and solution of carbon in austenite, the grain size is not always a true index to its stability. The high rate of transformation of small grains of nonuniform austenite may be attributed principally to depletion in carbon of the austenite at the instant of quenching rather than to the actual size of the austenite grains.

With the exception of a few of the grains of smaller size whose transformation on quenching was completed at a high temperature, each grain of austenite was revealed by each of its decomposition products, cementite and troostite.

The temperature at the start of the Arⁿ transformation progressively decreased with increase in carbon content of the alloys from about 900 degrees Fahr. for the 0.23 per cent carbon to about 500 degrees Fahr. for the 0.80 per cent carbon alloy.

Specimens of the hypereutectoid alloys, quenched from high temperatures at rates resulting in complete hardening, often showed numerous fine cracks located not only in the original austenitic grain boundaries but also in the interior of these grains. Intergranular cracks with respect to the original austenite were produced in these martensitic specimens by applying sufficient external stress to the quenched specimen to cause rupture.

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 6. H. Esser, W. Ellender and E. Spenle; "Quenching Diagrams of the Iron-Carbon Alloys," *Arch. Eisenhüttenw.*, Vol. 6, 1933, p. 389-393.
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DISCUSSION

Written Discussion: By H. W. McQuaid, assistant chief metallurgist, Republic Steel Corp., Cleveland.

It is interesting to note that Mr. Digges has shown that in high purity iron-carbon alloys, at least, the transformation rate is variable within the grain itself. He has shown that the distribution of carbon in the austenite at the time of quenching is an important factor in hardenability.

The use of the micro hardness test promises to help us greatly in our hardenability studies as we are able to visualize more easily the relation between the various structures shown and their hardness. The standard Rockwell test, valuable as it is as a commercial hardness tester, does not indicate the degree of transformation which has taken place, nor does it give any evidence of the maximum and minimum micro hardness in the area under the diamond impression.

The effect of carbon diffusion on hardenability can be demonstrated by varying the time and temperature of heating of steels of the same grain size but differing in their melting practice and final additions. Thus combinations of aluminum, nitrogen, oxides, etc. in the crystal and grain boundaries apparently increase greatly the resistance to carbon diffusion at the grain boundaries with resultant high transformation boundaries which result in decreased hardenability. This is a common method for decreasing hardenability in commercial steels where deep hardening is to be avoided. On the contrary where deep hardening is desirable it is important that the very minimum of boundary

constituents be present which interfere with carbon diffusion. Thus not only will a homogeneous distribution of carbon in the austenite be obtained but a homogeneous martensite will result which does not have boundaries of relatively low strength as compared to the martensitic grains. Such boundaries in the martensite reduce its ability to distribute stresses uniformly so that it is difficult when they are present to obtain high ductility in combination with high hardness.

Written Discussion: By Frederick C. Hull, Westinghouse Fellow in Metallurgy, Carnegie Institute of Technology and George E. Pellissier, Jr., International Nickel Co., Fellow in Metallurgy, Carnegie Institute of Technology, Pittsburgh.

The rates of transformation of commercial plain carbon steels are certainly influenced by several variables. The effects of carbon content and grain size have been investigated. Results are occasionally obtained on heat treating these steels which cannot be adequately accounted for by these factors alone. There is reason to believe, therefore, that other incidental factors, such as nonmetallic inclusions, gas content, amount of residual elements, the presence of undissolved cementite, and nonuniform distribution of carbon in austenite, may affect the rates of transformation. Until the importance of each of these factors has been evaluated, our understanding and control of the transformation process necessarily remain incomplete. It is obviously difficult to isolate the effects of the separate variables in investigations employing commercial steels. A more satisfactory method for evaluating the importance of these factors, incidental to steel making and heat treating practices, would involve studies of austenite decomposition in high purity iron-carbon alloys and the effects produced by adding these variables separately and then collectively. Mr. Digges has appreciated the importance of such studies as evidenced by his recent contributions.

In these investigations, isothermal transformation studies would indeed facilitate the interpretation of experimental results and are therefore to be preferred to quenching experiments. As the author indicates, such investigations are not easy to conduct, because of the relatively high rate of transformation of austenite in these high purity alloys. Nevertheless, it is believed that much of the desired information concerning the transformation process might be obtained from this type of investigation even though it is limited to subcritical temperatures above the so-called knee of the S-curve.

It appears to the writers that much of the confusion in understanding to which Mr. Digges refers is a result of a misinterpretation of the observed structures. An understanding of the rates of transformation of austenite is facilitated if the mechanism of the transformation is considered. The factors which directly determine the rates of transformation to pearlite are the rate of nucleation (N), the rate of growth (G), and the austenitic grain size (a), as described in detail by Mehl.¹ Although Mehl's analysis of the rates of decomposition of austenite was confined to steels of eutectoid composition, the basic principles set forth may be applied to studies of steels of other compositions. The incidental factors mentioned above can affect the rates of transformation only by influencing one or more of these fundamental variables, (N), (G), and (a). In the course of our studies we have found it conducive to a

¹R. F. Mehl, "The Physics of Hardenability," Hardenability of Alloy Steels Symposium, American Society for Metals, 1938, p. 1.

better understanding of the transformation process to analyze reaction rates in terms of these variables (N), (G), and (a), rather than to express the overall reaction rate in terms of the "relative stability" of austenite. In discussing Fig. 2, the author states: "The small unstable grains contained a higher percentage of troostite than the more stable large grains," thus inferring that small grains are inherently less stable than large grains and perhaps suggesting to the unwary reader that there is an instability about small grains wholly apart from the greater ratio of grain boundary area to grain volume. This is not the case, so far as we know, and such an assumption is unnecessary. Small grains complete their reaction more quickly than large grains because the pearlite, growing from nuclei formed at the grain boundary, has a shorter radial distance to grow. This is true when the rate of nucleation at the grain boundary and the rate of growth are identical in the two cases, as pointed out by Mehl.³ The effect is purely geometrical. The discussion at the bottom of page 578 and top of page 580 seems to disregard this; the greater percentage reaction of the smaller grains is a fact, but the amount of fine pearlite *per unit grain boundary area* is approximately the same in the two cases (actually somewhat less in the smaller grains owing to the greater curvature). This is not a new point of view (cf. Mehl); the reasoning is simple and quantitative; surely it is preferable to bulking all effects into ill-defined terms such as stability and instability which tend to disregard the elements which determine rates.

The author's observation that nuclei form preferentially at grain boundaries in these high purity alloys, as well as in commercial steels, is interesting, for it is evidence for the point of view that nucleation at grain boundaries does not require the presence there of inclusions or impurities. Atoms in the vicinity of a grain boundary should possess a higher average energy, because of lattice disregistry and accompanying strain, than atoms within the grain. This furnishes an explanation of preferential grain boundary nucleation of pearlite, for the additional energy required to form a stable nucleus is less for those atoms at the grain boundary. In other words, there is a greater probability of a nucleus forming at a boundary than within a grain. This does not preclude, however, the formation of an occasional nucleus within a grain (Figs. 9, 10, 11).

The preferential formation of nuclei at a proeutectoid cementite-austenite interface might be explained on the basis of a treatment by Mehl and Jetter⁴ which indicates that a nucleus forming at an interface requires less energy than one forming within a phase. This does not require that even a majority of the proeutectoid cementite particles act as centers of pearlite nucleation, since the formation of nuclei at these locations is a matter of probability. It is therefore unnecessary to assume, as the author has in discussing Fig. 12, that the role of proeutectoid cementite in nucleating pearlite is dependent on certain undefined "conditions."

An observation presented by Mr. Digges to demonstrate that undissolved carbides may act as centers for pearlite nucleation is difficult to interpret. One may not conclude from Fig. 13 that "some of the undissolved carbide particles or free carbides acted as nuclei for transformation during cooling," for it was

³R. F. Mehl, loc. cit., p. 12-14.

⁴R. F. Mehl and L. K. Jetter, "The Mechanism of Precipitation from Solid Solution," Age Hardening of Metals Symposium, American Society for Metals, 1939, p. 342.

impossible for nodules of the size shown to form without enveloping some carbide particles. He has neglected to mention that these undissolved carbides might influence the rate of transformation indirectly through their effect on the grain size and the carbon content of the austenite. The increased reaction rate noted cannot in any case be wholly attributed to an increased rate of nucleation of pearlite by the carbides.

There is nothing unusual in the observation that, in a steel partially reacted at a temperature slightly below the critical, a nucleus of pearlite should form in a large grain and not in the adjacent small grain (Figs. 6C and 8A). In fact, it is more likely that a nucleus will form in a *given* large grain than in a *given* small grain. The formation of nuclei is purely a probability effect; on the average the same number of nuclei will form on each unit area of grain boundary, but their distribution is statistical. At high transformation temperatures, the rate of nucleation is extremely low. Actually, a given grain will often react completely from a single nucleus. After partial transformation, there are still many grains in which no nuclei have formed. Since nuclei are statistically distributed, it is not to be expected that they will be found preferentially in either small or large grains. It is unnecessary to assume and difficult to imagine how carbon segregation prior to quenching could have been responsible, even in part, for the observed structure.

The assumption, of the existence of a carbon gradient in the austenite to explain the range of decomposition products noted in the large grain of Figs. 9, 10, 11, is entirely unwarranted. Proeutectoid cementite was formed at the grain boundary and in a Widmanstätten pattern within the grain during cooling from 1700 to 1330-1310 degrees Fahr. At the latter temperature, the abnormality, that is the free ferrite associated with the proeutectoid cementite, developed. The carbon in the austenite adjacent to the cementite diffused to the already existing cementite; this austenite depleted in carbon transformed to ferrite. Simultaneous with this process, coarse pearlite was nucleated at a few locations in the vicinity of the grain boundary and grew into the grain. The time of holding at this temperature was insufficient, however, to result in complete transformation of the grain to coarse pearlite. On quenching from 1330 to 1310 degrees Fahr., the remaining austenite transformed partially to fine pearlite at temperatures corresponding to the knee of the S-curve. The fine pearlite was nucleated preferentially at the ferrite-austenite interface and grew radially toward the center of the grain. The uniform band of fine pearlite observed was caused by the high rate of nucleation at this interface. The growth of fine pearlite was interrupted by the transformation to martensite at the lower temperatures prevailing at the end of the quench. The appearance of a nodule of fine pearlite within the grain has already been discussed.

The phenomena observed by the author can be understood without having to resort to the assumption that the distribution of carbon in the austenite was nonuniform. The term "nonuniform austenite," used to denote both incomplete solution and nonuniform distribution of carbon in austenite, is unfortunate. The effects of these two types of nonuniformity should receive separate treatment. From the work of the author and others, the influence of proeutectoid cementite is evident. It is conceivable that a nonuniform distribution of carbon

may exist in the austenite prior to quenching. However, the existence and importance of such a condition have not been demonstrated in Mr. Digges' experiments.

Written Discussion: By Howard Scott, section engineer, metallurgical section, Westinghouse Electric & Mfg. Co., E. Pittsburgh, Pa.

Mr. Digges has determined the start of the Ar" transformation in pure iron-carbon alloys under conditions which presumably assure that all carbon is retained in solid solution. This is a very difficult determination to make because of the extremely fast cooling rate necessary to retain the carbon in solid solution. Because of this fact, previously available data was not very consistent as shown by Scott and Hoop, Vol. 21 of Transactions, Fig. 12. Recomparison is now desirable to see how well the location of this transformation is established.

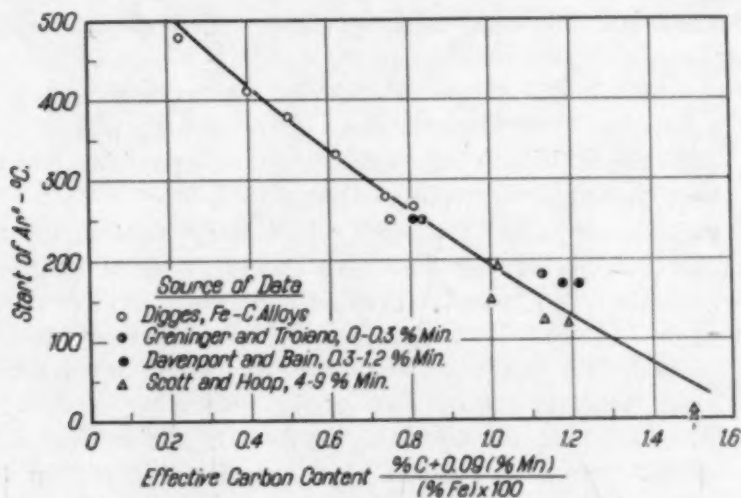


Fig. A—Start of Ar" in Iron-Manganese-Carbon Alloys so Cooled as to Retain all Carbon in Solid Solution.

Fig. A provides the comparison suggested. Digges' data is consistent with itself as well as with that of Greninger and Troiano⁴ and of Davenport and Bain.⁵ In the latter references, values of Ar" are not given but may be deduced from the data recorded. Both quench into baths at various temperatures thereby avoiding the need for observing temperatures during very rapid cooling as Digges does by an undisclosed means.

The other comparative data were recalculated from data on alloys containing 4 to 9 per cent manganese given by Scott and Hoop. Their values extrapolated to zero manganese content by the device illustrated in Fig. A agree fairly well with that from the other sources. In fact not one observation falls over 40 degrees Cent. from the smooth curve drawn to represent all of the data. We have therefore a well established relation between the start of Ar" and solid solution carbon content in carbon and carbon-manganese steels.

⁴A. B. Greninger and A. R. Troiano, "Kinetics of the Austenite to Martensite Transformation in Steel," TRANSACTIONS, American Society for Metals, Vol. 28, 1940, p. 537.

⁵Transactions, American Institute of Mining and Metallurgical Engineers, Iron and Steel Division, 1930, p. 117.

Author's Reply

The author is appreciative of the discussion of this paper. Mr. Scott has assembled the present data on the start of the Ar'' transformation in high purity iron-carbon alloys with that of other investigators for alloys of high carbon content with and without appreciable proportions of manganese. He shows that the relation of the temperature of the start of Ar'' to effective carbon content is represented by a smooth curve for cooling rates sufficiently fast to retain all carbon in solution. The results of tests recently completed by the author show there is a wide range in the rate of cooling without an appreciable change in the temperature of the start of Ar'' in the high purity iron-carbon alloy containing 0.80 per cent carbon. For observed cooling rates ranging from 550 to 2600 degrees Fahr. per second, the start of Ar'' occurred within the range of about 450 to 500 degrees Fahr. A split transformation, with the start of Ar' at 1040 degrees Fahr. and Ar'' at 500 degrees Fahr., occurred in the alloy quenched at a rate of 550 degrees Fahr. per second. For these tests, the specimens of the 0.80 per cent carbon alloy were quenched directly from 1425 degrees Fahr. in hydrogen and the average cooling rate through the temperature range of 1110 to 930 degrees Fahr. was determined by means of the string galvanometer apparatus referred to in the text and described in detail by French and Klopsch (7).^a

Messrs. Hull and Pellissier are of the opinion that the factors that directly determine the rates of transformation of austenite to pearlite are the rate of nucleation (N), the rate of growth (G), and austenitic grain size (a) as described by Mehl (2). The author concurs in that a consideration of these factors is essential to a clear picture of the mechanism of this transformation. However, the difficulty of rendering a satisfactory explanation of all reactions involved in the transformation of austenite to pearlite in terms of these factors is recognized by Mehl (2) as is evident from the paper which has been used primarily as the basis of the discussion by his associates. Mehl stated that "In a pure carbon steel grain size is the only variable affecting the position of the curve" (that is, the S-curve), and in his closure to the discussion, "It must be admitted that the decrease in the rate of transformation with increasing carbon content is difficult to explain either on the basis of the effect of increasing carbon on the rate of nucleation or on the rate of growth; it is a basic point worthy of much thought." Furthermore, Messrs. Hull and Pellissier have also failed to mention the difficulties of determining (N) and (G) experimentally, the deficiency of accurate quantitative data for these two variables for steels differing widely in chemical composition and for steels that show general nucleation, i.e., steels in which nucleation is not confined to the grain boundaries.

Messrs. Hull and Pellissier have commented on the transformation of austenite of mixed grain size. Possibly it should have been mentioned in the text that the parent austenitic grains were viewed on a flat surface and because of their shape some of the grains that appear small on the polished surface in reality may be large grains cut near their boundaries. It should be

^aThe figures appearing in parentheses refer to the bibliography appended to the paper.

obvious that if the nucleation of austenite to pearlite was confined to the grain boundaries of the parent austenite the nodules of pearlite in the small grains had a shorter radial distance to grow to complete their transformation than did the large grains. However, with a steel of mixed austenitic grain size, the transformation to pearlite does not necessarily start at the same instant in the small and large grains. In discussing the effects of mixed grain size on the transformation rate, Mehl (2), on page 14 of the paper cited by Messrs. Hull and Pellissier, stated "The small grains will have reacted before the large ones begin to react and there will be a hump in the reaction curve as a result." This statement would indicate that small austenitic grains are in-

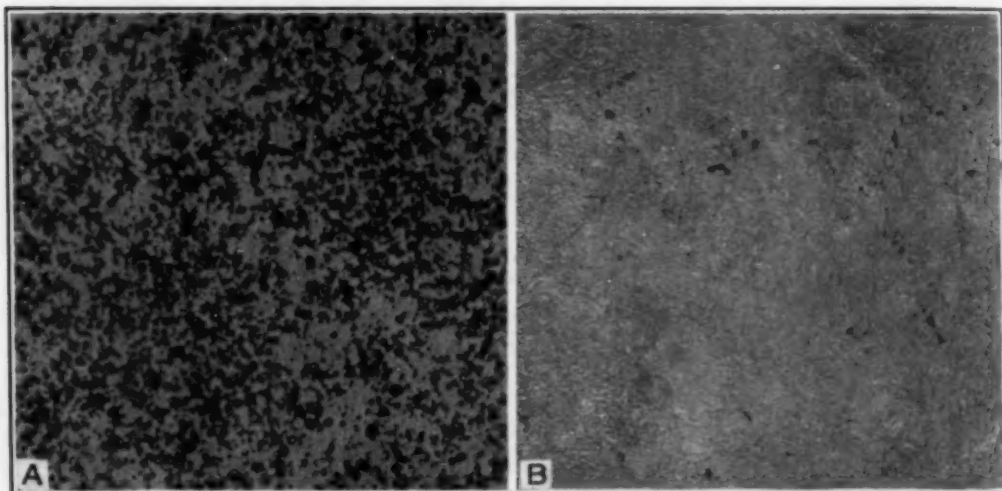


Fig. 1—Structure of Quenched Iron-Carbon Alloys. Specimens 0.10 Inch Square by 0.40 Inch Thick were Quenched Directly from 1425 Degrees Fahr. in Hydrogen.

A—Specimen Containing 1.01 Per Cent Carbon was Cooled Through the Temperature Range 1110 to 930 Degrees Fahr. at the Average Rate of Approximately 950 Degrees Fahr. per Second. The Structure of this Same Specimen is Also Shown in Fig. 13 of the Text. Etched with 1 Per Cent Nitric Acid in Alcohol. $\times 100$

B—Specimen Containing 0.80 Per Cent Carbon was Cooled Through the Temperature Range 1110 to 930 Degrees Fahr. at the Average Rate of Approximately 820 Degrees Fahr. per Second. Etched with 1 Per Cent Nitric Acid in Alcohol. $\times 100$.

herently less stable than large austenitic grains, a fact now generally recognized by most metallurgists.

It appears also that Messrs. Hull and Pellissier are of the opinion that the presence of fine carbide particles in austenite at the time of quenching has no appreciable influence on the transformation rate of austenite to pearlite. It must be admitted that it was impossible for nodules of troostite of the size of the majority shown in Fig. 13 to form without enveloping some of the carbide particles. This condition does not invalidate the conclusion made by the author that the transformation rate of the austenite in this specimen was influenced by the presence of free carbide. Possibly the data that follow should have been included in the test in support of this statement. The structure of this same specimen containing 1.01 per cent carbon is shown in Fig. 1A of this discussion and the structure of a specimen of a high purity iron-carbon alloy containing 0.80 per cent carbon is shown in Fig. 1B. Since both specimens were quenched directly from 1425 degrees Fahr., approximately the same amount of carbon was

in solution in each specimen at the time of quenching. At 1425 degrees Fahr., the 0.80 per cent carbon specimen had an average grain size of about 20 grains per square inch at 100 diameters (A. S. T. M. grain No. 5). Although there is sufficient troostite in the quenched specimen containing 1.01 per cent carbon, as shown in Fig. 1A, the troostite did not clearly delineate the parent austenitic grains. This condition was brought about by the presence of free carbide particles at the time of quenching. Apparently, the free carbides have caused a general nucleation and have rendered the specimen effectively fine-grained. If nucleation was confined to the austenitic grain boundaries, it is obvious that the specimen was very fine-grained, a condition which is unusual for high-purity iron-carbon alloys. The specimen containing 1.01 per cent carbon (Fig. 1A) was cooled through the temperature range of 1110 to 930 degrees Fahr. at the observed average rate of about 950 degrees Fahr. per second, whereas the specimen containing 0.80 per cent carbon (Fig. 1B) was cooled through the same temperature range at about 820 degrees Fahr. per second. The former specimen, although cooled at the fastest rate, nevertheless contained appreciably more troostite than the latter specimen. The free carbide particles at the time of quenching the specimen containing 1.01 per cent carbon evidently were directly responsible for its relatively high transformation rate.

Messrs. Hull and Pellissier state that "The assumption, of the existence of a carbon gradient in the austenite to explain the range of decomposition products noted in the large grains of Figs. 9, 10, 11, is entirely unwarranted." In their attempt to explain the transformation in this same grain they also state, "The carbon in the austenite adjacent to the cementite diffused to the already existing cementite; this austenite depleted in carbon transformed to ferrite." Thus, in their attempt to prove their statement that the assumption of a carbon gradient in the austenite is unnecessary to explain the decomposition products observed, the discussers immediately contradict themselves by stating that diffusion, with consequent depletion of carbon in the austenite, occurred.

Mr. McQuaid has commented on the use of grain boundary constituents for controlling the hardenability of commercial steels.

THE EFFECT OF TITANIUM ON THE MACROSTRUCTURE AND GRAIN COARSENING TEMPERATURE OF FORGING STEEL

BY GEORGE F. COMSTOCK

Abstract

Small ingots of steel containing 0.32 to 0.41 per cent carbon, 0.62 to 0.72 per cent manganese, 0.10 to 0.37 per cent silicon, and low phosphorus and sulphur, were poured at controlled temperatures from induction furnace melts to which aluminum, aluminum plus titanium, or aluminum plus zirconium, were added in various amounts just before pouring. The ingots were split lengthwise, one-half of each being etched by Humfrey's method to show the macrostructure, while the other half was forged to a 1-inch square bar. Samples from this bar were heated for 8 hours at temperatures from 1700 to 1950 degrees Fahr. respectively, and the grain sizes were determined after heating. Residual titanium contents of about 0.1 per cent or more were found to prevent the formation of coarse dendrites in these ingots, while with similar amounts of aluminum or zirconium but no titanium coarse dendrites were formed. With titanium contents above 0.2 per cent and aluminum additions of 0.05 to 0.15 per cent the grain coarsening temperature of this steel was found to be above 1950 degrees Fahr.; and titanium contents of 0.10 to 0.12 per cent with aluminum additions of 0.05 to 0.15 per cent gave higher grain coarsening temperatures than were obtained with about 0.1 per cent zirconium or by increasing the aluminum addition alone from 0.10 to 0.15 per cent.

THE first objective of the work described herein was to show how the grain coarsening temperature of a plain carbon steel of about 0.37 per cent carbon was affected by titanium contents between about 0.1 and 0.4 per cent together with aluminum in various small amounts. It was already known that a steel of this kind with 0.33 per cent titanium had a high grain coarsening temperature while one with 0.04 per cent titanium did not, both having high aluminum contents, and it was desired to determine what the effects of intermediate titanium contents would be, with aluminum contents not abnormally high. At the same time the macrostructures of the ingots made for

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. The author, George F. Comstock, is metallurgist, The Titanium Alloy Manufacturing Co., Niagara Falls, N. Y. Manuscript received April 26, 1939.

this investigation were examined to check the elimination of dendrites by titanium as was reported by Valenta and Poboril of the Skoda Works in Czecho-Slovakia in a paper presented in 1935 at the International Foundry Congress in Brussels, Belgium.

The results reported in this paper were obtained from steel melted in a small induction furnace with a magnesia crucible, each heat making a single ingot weighing about 17 pounds. The analyses, pouring temperatures, titanium and aluminum additions, etc., are given in Table I. Aluminum was added in the form of shot about 3 or 4 minutes before pouring, and the temperature was measured at that time, using an optical pyrometer sighted at the top of the bath. The readings were corrected for slag emissivity. The slag was removed as much as possible before adding titanium, which was in the form of about $\frac{3}{4}$ -inch lumps of low carbon ferrotitanium containing about 40 per cent titanium, 0.1 per cent carbon, 7 per cent aluminum, and 3.5 per cent silicon. The steel was poured 1 or 2 minutes after this addition, which dissolved rapidly. Two heats received no titanium, and for comparison two others were investigated in which a ferro alloy containing about 35 per cent zirconium, 24 per cent aluminum, and 5 per cent silicon was substituted for the ferrotitanium. The phosphorus and sulphur in all these steels were below 0.04 per cent. The aluminum was determined on only a few samples, and the values reported in Table I represent average results for total aluminum by two accurate chemical methods.

The ingots cast from the heats listed in Table I were split in half lengthwise, one-half of each being machined flat, polished and

Table I
Analyses, Additions, and Pouring Temperatures of Steels Tested

Heat No.	Chemical Analysis					Titanium		Aluminum	Pouring Temperature °F.
	C	Mn	Si	Al	Ti	Per Cent Added	Per Cent Recovery	Per Cent Added	
147	0.38	0.63	0.19	None	..	0.10	2965
149	0.38	0.72	0.37	0.13	None	..	0.15	2925
156	0.41	0.66	0.18	0.105	0.16	66	0.05	2925
155	0.38	0.67	0.23	0.105	0.16	66	0.10	2925
154	0.34	0.62	0.12	0.12	0.16	75	0.15	2890
151	0.32	0.64	0.27	0.15	0.33	46	0.15	2940
153	0.35	0.64	0.19	0.24	0.33	73	0.05	2940
152	0.37	0.65	0.31	0.24	0.33	73	0.10	2910
159	0.40	0.64	0.16	0.11	0.27	0.33	82	0.15	2910
161	0.37	0.71	0.28	0.30	0.49	61	0.05	2890
157	0.41	0.66	0.10	0.375	0.49	76	0.05	2940
158	0.40	0.65	0.19	0.10	0.39	0.49	80	0.10	2910
162	0.40	0.67	0.24	0.05	(0.036 Zr	0.05 Zr	72)	0.10	2930
163	0.40	0.67	0.14	(0.098 Zr	0.11 Zr	89)	0.075	2890

etched to show the macrostructure, while the other half was forged into a bar about 1-inch square. The polished half-ingots were etched by Humfrey's method¹ to show the dendritic structure. This method involves etching first with neutral copper-ammonium chloride solution, then replacing it with an acid solution, and after washing and drying, rubbing the etched surface gently with a fine abrasive paper. The macrostructures as revealed by this treatment are indicated in Table II.

Table II
Macrostructures of Lengthwise Ingot Sections

Heat No.	Titanium Content %	Pouring Temp. °F.	Appearance after Etching by Humfrey's Method
147	2965	Distinct coarse dendrites.
149	2925	Distinct dendrites, longer.
156	0.105	2925	Only fine traces of dendrites.
155	0.105	2925	Coarse specks, no dendrites.
154	0.12	2890	Very fine indistinct traces of dendrites.
151	0.15	2940	Very fine dendrites, slightly more distinct.
153	0.24	2940	No dendrites, very fine specks.
152	0.24	2910	Like 154 except finer.
159	0.27	2910	Same as 152.
161	0.30	2890	Fine traces of dendrites, like 156.
157	0.375	2940	Coarse specks, no dendrites, like 155.
158	0.39	2910	Same as 157.
162	(0.036 Zr)	2930	Coarse dendrites, rather long.
163	(0.098 Zr)	2890	Coarse but shorter dendrites.

Six photographs of typical etched sections, enlarged about 2.5 times, are presented as Figs. 1 to 6 to illustrate the results reported in Table II. It is evident that there is a marked difference between the titanium and titanium-free steels, none of the former showing noticeable dendrites, while in all the latter they were very distinct. A titanium content of only 0.105 per cent, which was the lowest amount tried, seems to be sufficient for dendrite elimination in these small ingots of 0.35 to 0.40 per cent carbon steel. The amounts reported as being effective by Valenta and Poboril were 0.26 to 0.31 per cent in larger ingots of about the same kind of steel. A recent test in an American steel foundry showed that 0.024 per cent titanium was not sufficient to make much difference in the dendrites in large castings of 0.31 to 0.37 per cent carbon steel.

Some of the forged bars made from the other halves of the ingots were sectioned lengthwise about midway between the bottom and top ends, and etched by Humfrey's method as the ingots were. Since the reduction in forging amounted to only about 3 to 1 in

¹*Journal, Iron and Steel Institute*, Vol. 99, 1919, p. 273.

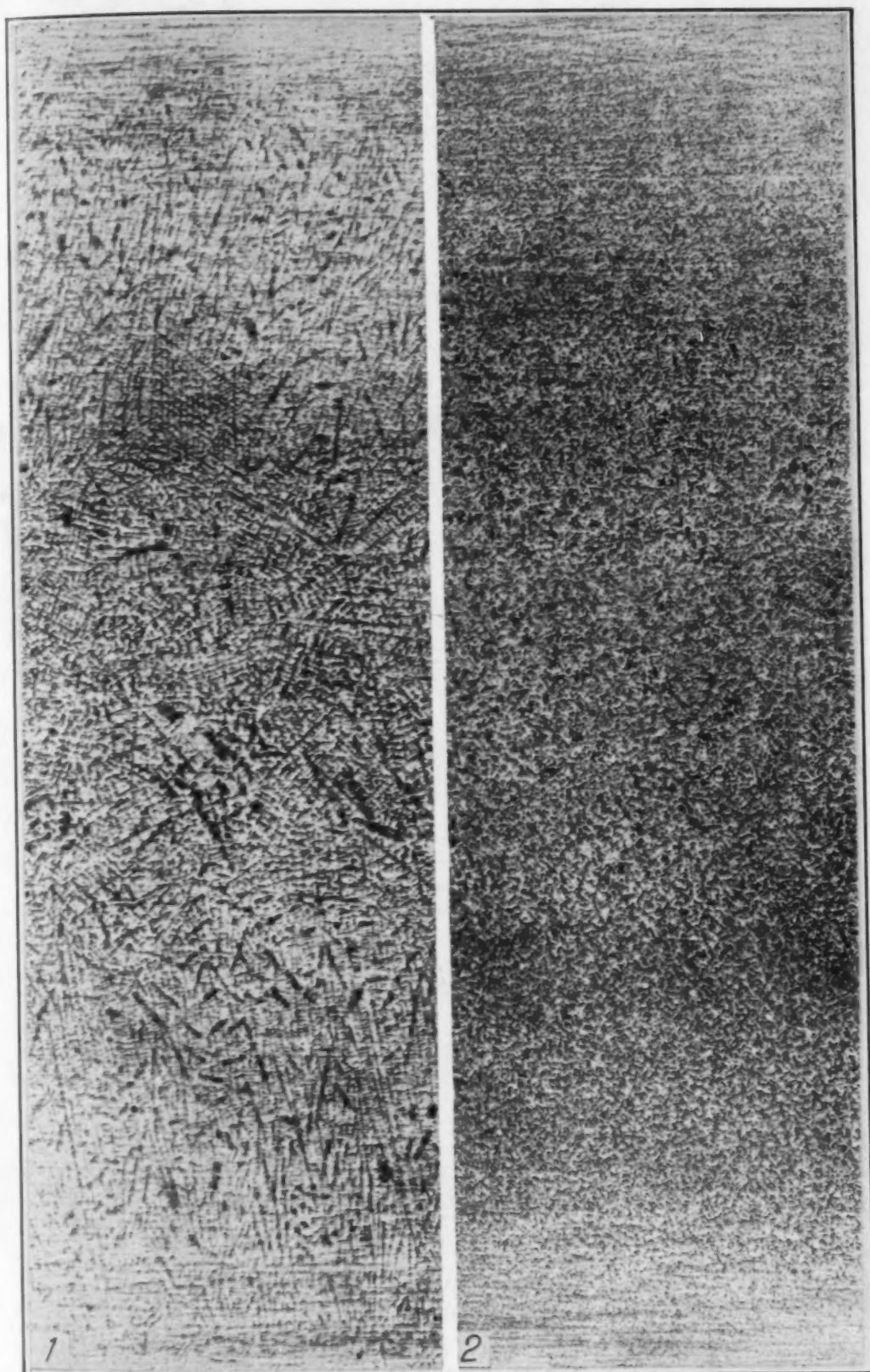


Fig. 1—Heat 149 Poured at 2925 Degrees Fahr., 0.15 Per Cent Aluminum Added. No Titanium Added.

Fig. 2—Heat 156 Poured at 2925 Degrees Fahr., 0.05 Per Cent Aluminum Added and Containing 0.105 Per Cent Titanium.

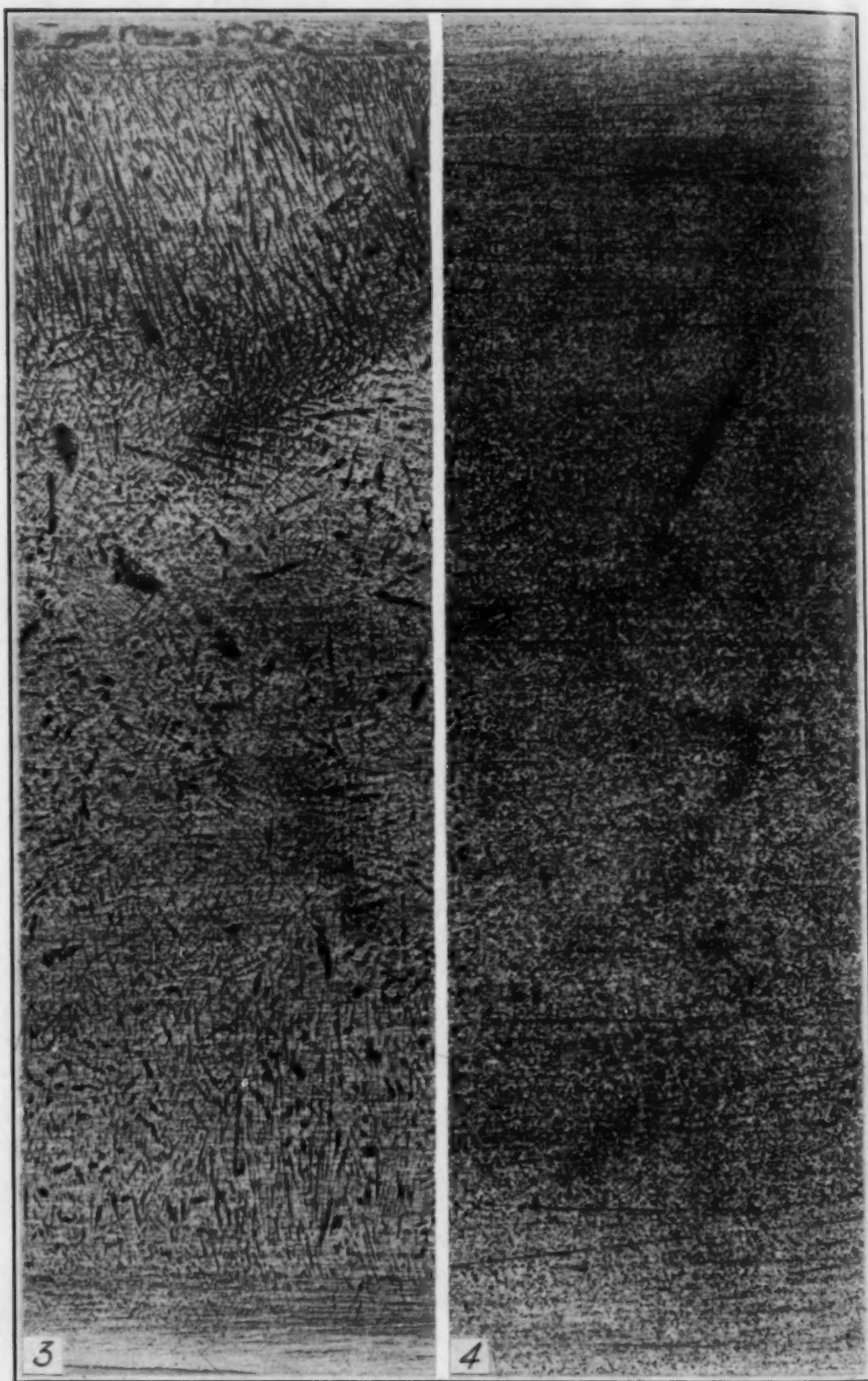


Fig. 3—Heat 162 Poured at 2930 Degrees Fahr., 0.1 Per Cent Aluminum and 0.05 Per Cent Zirconium Added. No Titanium.

Fig. 4—Heat 154 Poured at 2890 Degrees Fahr., 0.15 Per Cent Aluminum Added and Containing 0.12 Per Cent Titanium.

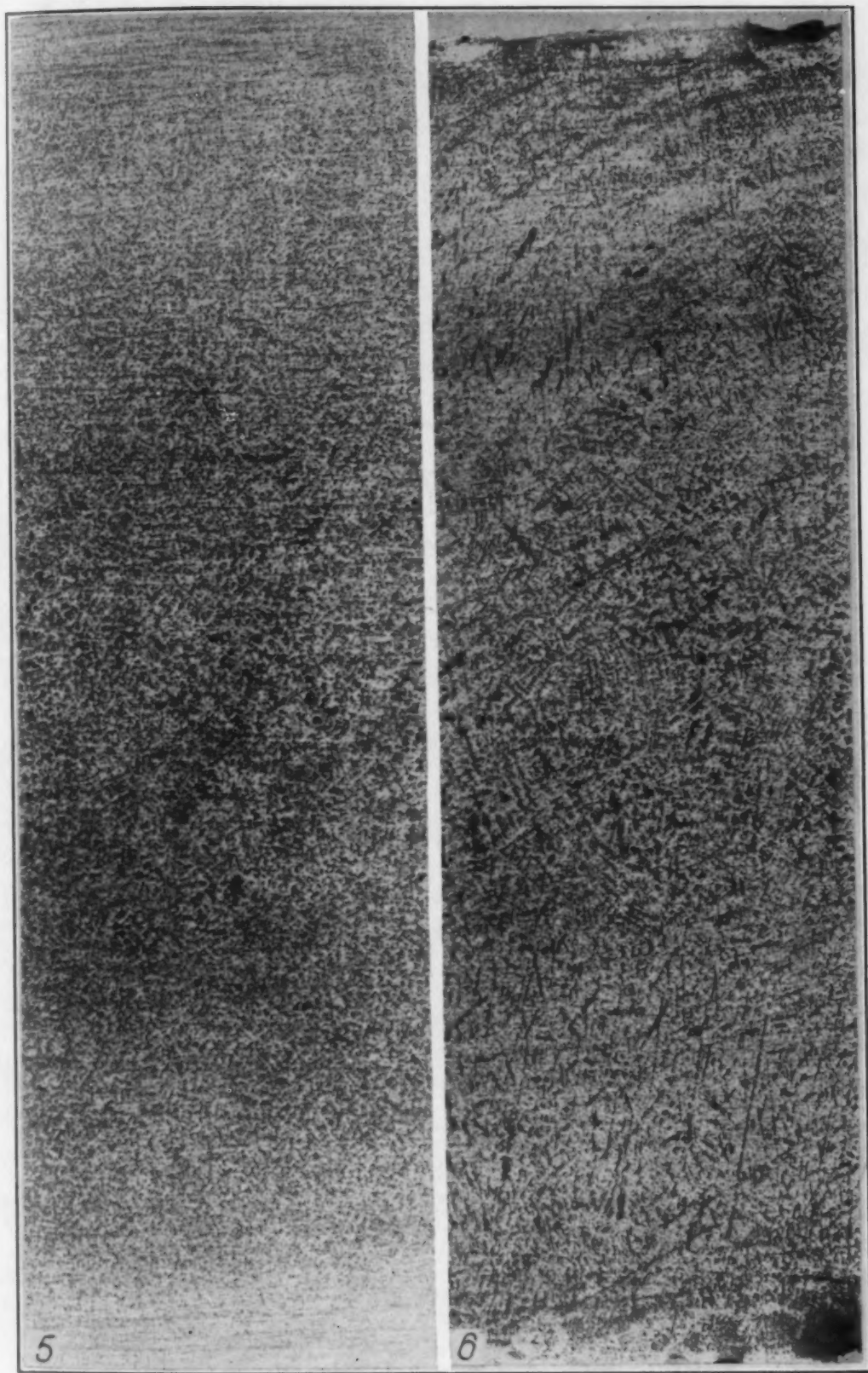


Fig. 5—Heat 161 Poured at 2890 Degrees Fahr., 0.05 Per Cent Aluminum Added and Containing 0.30 Per Cent Titanium.

Fig. 6—Heat 163 Poured at 2890 Degrees Fahr., 0.075 Per Cent Aluminum and 0.11 Per Cent Zirconium Added. No Titanium.

cross sectional area the dendrites in the dendritic steels were not transformed into streaks or "banding", but were merely distorted. The nondendritic titanium steels also showed about the same kind of etched surface on the bar sections as on the ingot sections, and it seemed unnecessary, therefore, to illustrate the etched sections of the bars.

The forged bars were sectioned crosswise just above a short bottom discard, and also about $8\frac{1}{2}$ inches from the bottom end, and their grain sizes without any heat treatment were determined by comparing etched sections at 100 diameters magnifications with the standard A.S.T.M. photomicrographs of hypoeutectoid grain sizes. All the steels were found to have grain sizes corresponding to No. 7 or finer, only one (No. 163) being no finer than No. 7. Pieces $1\frac{1}{2}$ inches long were cut from the bottom portions of the forged bars and heated for 8 hours at constant temperature, followed by cooling in the furnace. One piece from each steel was tested at 1700 degrees Fahr. (925 degrees Cent.), another set at 1750 degrees Fahr. (955 degrees Cent.), and so on up to 1900 degrees Fahr. (1040 degrees Cent.). Microsections were cut from the centers of these pieces after heating, and their grain sizes were determined in the way described above. The results are reported in Table III. Grain coarsening was considered to have occurred when any grains of No. 5 size or coarser were found in the section examined. Those steels which showed no coarsening at 1900 degrees Fahr. (1040 degrees Cent.) were tested further with additional pieces at 1950 degrees Fahr. (1065 degrees Cent.). Even this was not high enough to determine the grain coarsening temperature of the higher titanium

Table III
Grain Sizes of Forged Bars Annealed 8 Hours

Heat No.	Titanium Content Per Cent	Aluminum Added Per Cent	Annealing Temperature, °F.						Highest Temperature before Grain Coarsening
			1700	1750	1800	1850	1900	1950	
147	0.10	8	8	7-8	7-8	3-7	...	1850
149	0.15	7-8	7-8	7-8	4-8	3-7	...	1800
156	0.105	0.05	7-8	7-8	7	7	3-7	...	1850
155	0.105	0.10	8	7	7	6-7	4-7	...	1850
154	0.12	0.15	7	7	7	7	6	5	1900
151	0.15	0.15	7-8	7-8	5-7	4-7	4-7	...	1750
153	0.24	0.05	7-8	7-8	7-8	7	7	7	At least 1950
152	0.24	0.10	8	8	8	7-8	7	7-8	At least 1950
159	0.27	0.15	8	8	8	7-8	7	7	At least 1950
161	0.30	0.05	8	8	8	7-8	7	7	At least 1950
157	0.375	0.05	8	8	8	7-8	7	7	At least 1950
158	0.39	0.10	8	7-8	8	8	7	7	At least 1950
162	(0.036 Zr)	0.10	7-8	7-8	7	7	3-6	...	1850
163	(0.098 Zr)	0.075	6-7	3-7	2-7	2-6	2-6	...	1700

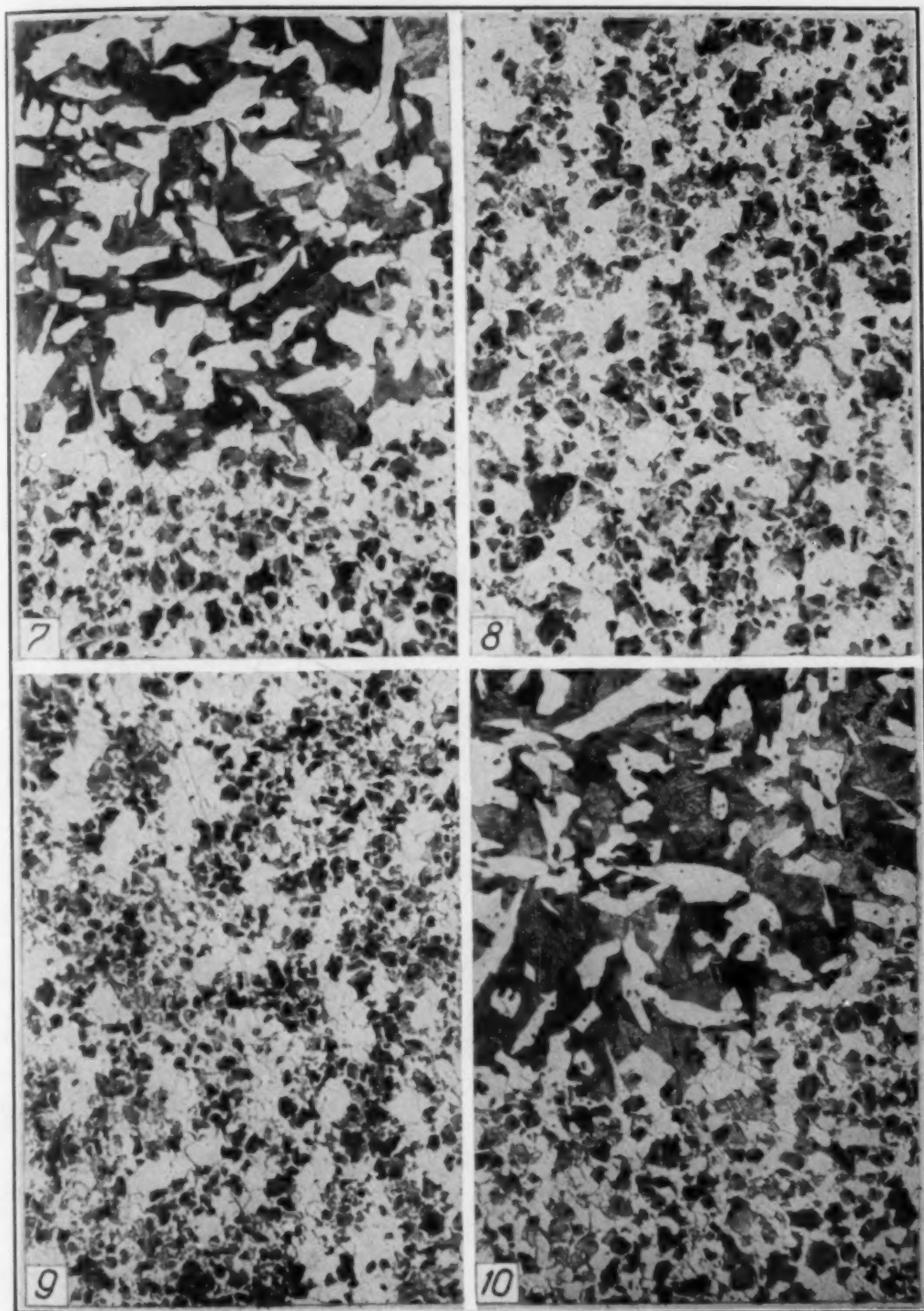


Fig. 7—Steel 149, with 0.15 Per Cent Aluminum Added and No Titanium, Heated 8 Hours at 1850 Degrees Fahr., Showing Grain Coarsening from 8 to 4.

Fig. 8—Steel 156, with 0.05 Per Cent Aluminum Added and 0.105 Per Cent Titanium Content, Heated 8 Hours at 1850 Degrees Fahr., Showing Uniform Grain Size 7.

Fig. 9—Steel 152, with 0.10 Per Cent Aluminum Added and 0.24 Per Cent Titanium Content, Heated 8 Hours at 1950 Degrees Fahr., Showing Grain Size of 7 to 8.

Fig. 10—Steel 163, with 0.075 Per Cent Aluminum Added and No Titanium But 0.098 Per Cent Zirconium Content, Heated 8 Hours at 1750 Degrees Fahr., Showing Grain Coarsening from 7 to 3.

All Show Cross Sections of 1-Inch Square Forged Bars, HNO_3 Etch. $\times 100$.

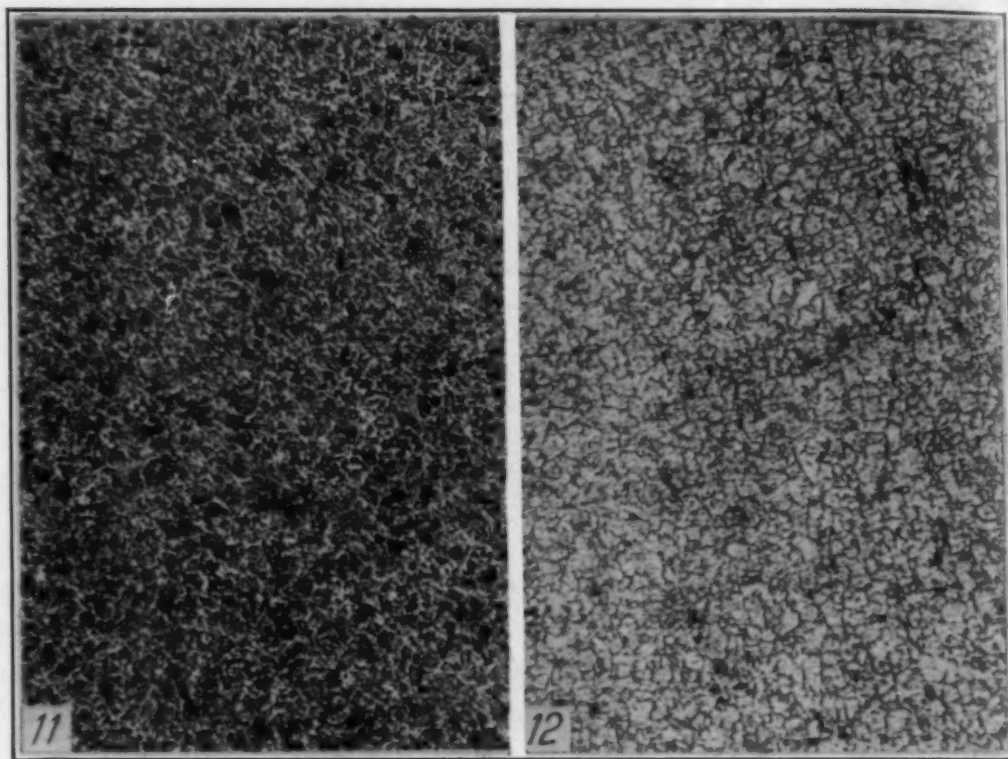


Fig. 11—Near Air-Cooled End Showing Sorbite Grains Outlined with Ferrite.
Fig. 12—Near Brine-Cooled End Showing Martensite Grains Outlined with Troostite.
Sections of Small Piece of Forged Steel 157, with 0.05 Per Cent Aluminum Added and 0.375 Per Cent Titanium Content, Heated 4 Hours at 1950 Degrees Fahr. and Cooled with Only One End in Brine; Etched with Nitric Acid and Magnified 100 Diameters, Showing Grain Size of 7 to 8.

steels, but the extension of the work to still higher temperatures was not thought to have sufficient practical value to justify the additional time and risk of damage to equipment which would be involved.

To check the grain size estimations of Table III, small pieces of the six steels reported as having No. 7 grain size after annealing at 1950 degrees Fahr. (1065 degrees Cent.) were heated 4 hours at the same temperature and tested by the gradient quench method described by Vilella and Bain.² One specimen was spoiled by falling lengthwise in the quench, but the other five showed grain sizes of 7 to 8 by this method, which was considered a satisfactory check on the simpler procedure used for obtaining the data reported in Table III.

Photomicrographs are presented as Figs. 7 to 12 to illustrate these grain sizes. The uneven distribution of pearlite in the fine-grained annealed steels was undoubtedly due to the small reduction

²J. R. Vilella and E. C. Bain; "Revealing the Austenitic Grain Size of Steel," *METAL PROGRESS*, Vol. 30, No. 3, September 1936, p. 39.

in hot working from the half-ingot to the forged bar, which as noted above was insufficient to break up the ingot structure completely. The uneven structure evidently did not affect the reliability of the grain size estimations, since the gradient quench method showed a similar fine grain after the highest heat treatment.

In Table III it may be noted that all the steels with over 0.2 per cent titanium had a grain coarsening temperature at least as high as 1950 degrees Fahr. (1065 degrees Cent.), while those with 0.1 to 0.2 per cent titanium had grain coarsening temperatures above 1850 degrees Fahr. (1010 degrees Cent.) with the exception of heat 151. This heat had the lowest carbon content of the series, and it was also exceptional in showing a low recovery of the titanium added. Perhaps there is some connection between these indications of over-oxidation and the low grain coarsening temperature, so that more reliance should be placed on the evidence from the other heats.

Variations in the amounts of aluminum added do not seem to have made much difference in these steels except that the larger additions tended to improve the recovery of titanium. Heat 149 shows that higher aluminum without titanium did not tend to raise the grain coarsening temperature. No evidence is afforded by the two heats containing zirconium that this element is helpful in maintaining fine grain at high temperature. A comparison of heat 163 with 155 and 156 certainly indicates that titanium was much more effective.

CONCLUSION

In conclusion it should be emphasized that the amounts of titanium discussed here are residual titanium contents, and not additions. The effects described, either on the dendrites or on the grain size, are not obtained by mere deoxidation with titanium, such as has been generally the objective in commercial use of titanium in forging steel. The residual titanium contents necessary for the elimination of dendrites or a marked increase in the grain coarsening temperature are more readily obtained with low carbon ferrotitanium than with high carbon, and preliminary deoxidation of the steel with aluminum is very helpful in securing satisfactory recovery of titanium. The results of this investigation may be summarized as follows:—

- (1) Residual titanium contents of about 0.1 per cent or more prevent the formation of coarse dendrites in small ingots of 0.35 to

0.40 per cent carbon steel poured at about 2900 to 2950 degrees Fahr. (1595 to 1620 degrees Cent.), while similar amounts of aluminum or zirconium were not effective.

(2) With over about 0.2 per cent residual titanium content and aluminum additions of 0.05 to 0.15 per cent the grain coarsening temperature of forged 0.35 to 0.40 per cent carbon steel was found to be above 1950 degrees Fahr. (1065 degrees Cent.).

(3) Residual titanium contents of 0.10 to 0.12 per cent with aluminum additions of 0.05 to 0.15 per cent gave higher grain coarsening temperatures in forged 0.35 to 0.40 per cent carbon steel than were obtained with about 0.1 per cent zirconium or by increasing the aluminum alone from 0.10 to 0.15 per cent.

(4) For the elimination of coarse dendrites, or raising the grain coarsening temperature, definite residual titanium contents in the steel are necessary and not mere deoxidation with titanium; and preliminary deoxidation with aluminum helps to retain the required titanium in the steel.

DISCUSSION

JEROME STRAUSS:^a I should like to add a word in confirmation of Mr. Comstock's findings. In a series of experiments on cast steels a short time ago we found very definitely an increase in the grain coarsening temperatures as a result of the addition of titanium and also that the addition, if made in the form of a single alloy containing aluminum and titanium along with silicon, was extremely effective. We also confirmed his findings in that the substitution of zirconium for titanium in the same alloy was useful but that the effect was much less pronounced.

Author's Reply

Mr. Strauss's confirmation of my findings is very much appreciated. Although there is no doubt but that good results can be obtained with the complex alloy that he mentions, it should be noted that my results were obtained by adding first silicon, then aluminum, and finally 40 per cent low-carbon ferro-titanium to the molten steel.

^aVice-president, Vanadium Corporation of America, Bridgeville, Pa.

THE NICKEL-TUNGSTEN SYSTEM

BY FINLEY H. ELLINGER AND WESLEY P. SYKES

Abstract

This system has been systematically studied by the methods of metallography and X-ray diffraction, supplemented by measurements of electrical resistivity and hardness. The constitutional diagram as constructed from the data thus obtained shows a maximum of 1505 degrees Cent. (2740 degrees Fahr.) in the solidus and liquidus at a tungsten concentration of about 35 per cent by weight. The solubility of tungsten in the nickel-rich solid solution decreases from 40 per cent at 1495 degrees Cent. (2725 degrees Fahr.), the eutectic temperature, to 38 per cent at 970 degrees Cent. (1780 degrees Fahr.). Between 970 and 800 degrees Cent. (1780 and 1470 degrees Fahr.) the solubility of tungsten drops to about 32 per cent. At 970 degrees Cent. a peritectoid reaction occurs by which is formed the intermediate phase with a tungsten concentration of approximately 43 per cent. This reaction involves the nickel-rich solid solution and the tungsten-rich solid solution, the second of which dissolves about 0.3 per cent nickel at the eutectic temperature. The eutectic occurs at a tungsten concentration of about 45 per cent.

Precipitation hardening was observed in the range of composition between 32 and 45 per cent tungsten. Hardening takes place very slowly at 600 degrees Cent. (1110 degrees Fahr.) and with increasing rapidity as the aging temperature is raised to 900 degrees Cent. (1650 degrees Fahr.). The tendency to overage even at 900 degrees Cent. (1650 degrees Fahr.) is very slight, as the maximum hardness of 490 Brinell developed in the 45 per cent tungsten alloy after 100 hours at 900 degrees Cent. (1650 degrees Fahr.) is reduced only some 20 points during a second 100-hour period at this aging temperature.

IN studying the age hardening properties of certain nickel-rich compositions in the nickel-tungsten system certain unusual features were observed. In order to arrive at a more thorough under-

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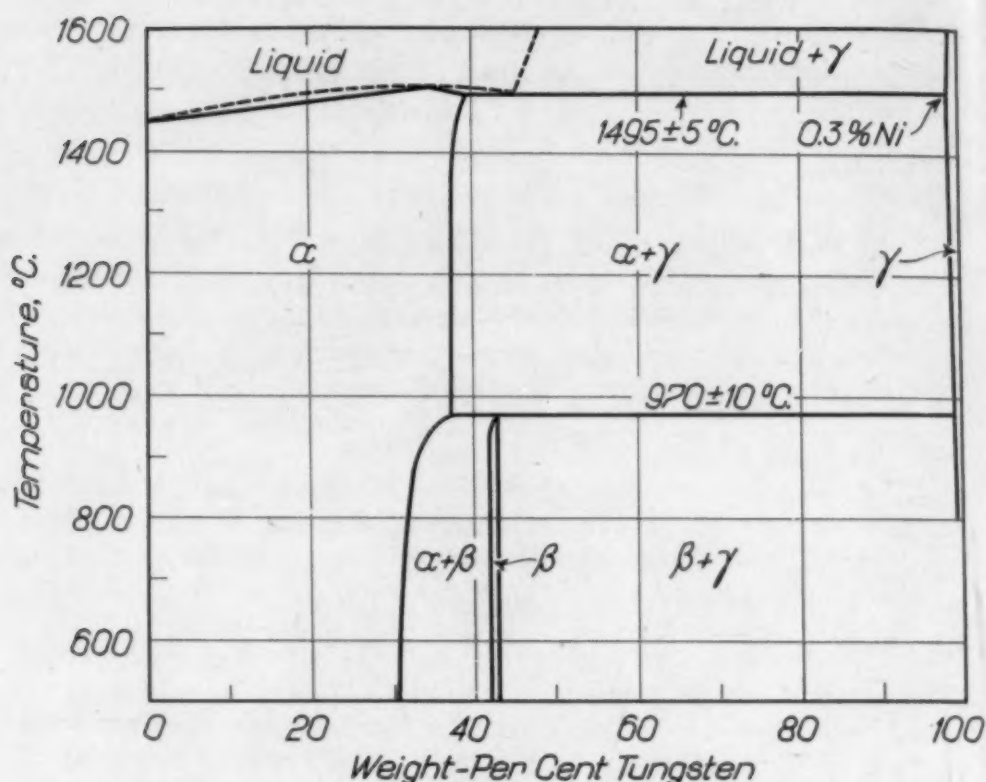


Fig. 1—Constitutional Diagram of Nickel-Tungsten System.

standing of these alloys the binary system was systematically explored by the methods of metallography and X-ray diffraction, supplemented by measurements of electrical resistivity and hardness.

Previous investigations of this system have been reported by R. Irrmann (1)¹ and by R. Vogel (2). Vogel placed the eutectic composition at 52 per cent tungsten and the temperature at 1510 degrees Cent. (2750 degrees Fahr.), the two solid phases in the eutectic structure being a nickel-rich solid solution and tungsten. Vogel's diagram indicated a eutectoid reaction at 920 degrees Cent. (1690 degrees Fahr.), whereby the nickel-rich solid solution of 47 per cent tungsten content decomposed to yield tungsten and a phase designated as Ni_6W , having a limiting composition on the tungsten-rich side at about 34 per cent tungsten.

Fig. 1 represents the constitutional diagram of this system drawn from the results of the present investigation. Little change has been made in the form or temperature range of the solidus as shown by Vogel, but no evidence could be observed to confirm the existence of the eutectoid reaction or the phase Ni_6W resulting from

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

it. On the other hand, the existence of a peritectoid reaction is clearly indicated, the temperature of which is placed at 970 ± 10 degrees Cent. This reaction appears to involve the nickel-rich solid solution (α) and the tungsten-rich solid solution (γ) yielding as a product the β phase which occurs at about 43 per cent tungsten. The homogeneity range of this phase cannot be greater than 2 per cent tungsten at a temperature of 900 degrees Cent. (1650 degrees Fahr.) but higher precision is not possible because of the very low diffusion rates characterizing these ranges of compositions and temperatures.

Hydrogen reduced metal powders were used in the preparation of the alloys. C. P. oxide was the source of nickel, while the tungsten powder contained not more than 0.1 per cent impurities. The powders in the desired proportions were mixed by tumbling, pressed into small bars and subsequently sintered or melted in a hydrogen atmosphere. To produce a homogeneous alloy by sintering it was found helpful to ball-mill the powders for some time before pressing. This procedure was especially effective in attaining complete solution as the tungsten concentration of the (α) solid solution approached the saturation limit of about 40 per cent at the solidus temperature. To determine the extent of possible contamination of the molten metal from the alundum boats used in melting the pressed powders, a chemical analysis was obtained from some of the nickel after melting and solidifying in hydrogen. This showed the presence of 0.37 per cent Fe, 0.10 per cent Si and 0.008 per cent Al, some of which may have been picked up from the alundum boat.

For most purposes of this investigation the alloys after melting were reduced 50 to 75 per cent in thickness by hot rolling. Alloys of tungsten content between 1 and 45 per cent were worked successfully at temperatures ranging between 900 and 1300 degrees Cent. (1650 and 2370 degrees Fahr.), the higher temperatures being necessary for the higher tungsten contents. Some difficulty, however, was encountered in working the alloys of 25 to 38 per cent tungsten due to cracking, chiefly at the grain boundaries. This difficulty appeared to arise from gas occluded at the grain boundaries during solidification of the metal, for it was difficult to produce sound, gas-free ingots by melting in hydrogen, especially in this range of compositions.

CONSTITUTIONAL DIAGRAM

Nickel and tungsten form a eutectiferous series of alloys as may

be seen in Fig. 1. Alloys up to 39 per cent tungsten solidify as alpha solid solution and from 39 to 45 per cent tungsten as primary alpha plus the eutectic. Fig. 2 shows the structure of the eutectic consisting of curved rod-like particles of the gamma phase embedded in a matrix of alpha solid solution. Increasing the tungsten content beyond 45 per cent results in the appearance of primary gamma dendrites, as in Fig. 3. With increasing tungsten concentration, the primary dendrites of (γ) tend to segregate in the lower part of the molten alloy.

The composition of the (γ) phase at the eutectic temperature was approximated by the addition of fractional percentages of nickel to tungsten oxide in the form of a solution of nickel acetate. After reduction, pressing and sintering for several hours at a temperature slightly below the solidus the microstructure showed both (α) and (γ) in the alloy containing 0.5 per cent nickel. When the nickel content was lowered to 0.2 per cent, however, the (γ) phase only was observed. The solid solubility of nickel in tungsten is, therefore, placed at about 0.3 per cent.

From pure nickel the solidus-liquidus temperatures rise gradually with increasing tungsten concentration up to a maximum of 1505 ± 5 degrees Cent. (2740 degrees Fahr.) at about 35 per cent tungsten. Between 30 and 45 per cent tungsten the melting range is very narrow and flat, the solidus temperature falling only 10 degrees Cent. from the maximum at 35 per cent tungsten to the eutectic temperature at 39 per cent tungsten. Beyond the eutectic point the liquidus appears to rise rapidly toward the melting point of tungsten.

To determine the solidus temperatures of the alloys up to 45 per cent tungsten, observations of fusions were made with an optical pyrometer. Specimens in the form of tall, slender cones were cut from rolled slabs of a series of compositions. Sets of three specimens of different compositions were mounted in an upright position on alundum blocks in the tube of the hydrogen furnace. For each determination a set of cones was placed in the furnace at 1400 to 1450 degrees Cent. (2550 to 2640 degrees Fahr.) and held at that temperature for an hour, after which the temperature was raised at the rate of 1 to 2 degrees Cent. per minute. The first indication of liquid appeared as a dark spot or line on the specimen which gradually sagged and finally fell into a ball of liquid.

The eutectic temperature as determined by the fusion of the 45 per cent tungsten alloy is placed at 1495 ± 5 degrees Cent. (2725

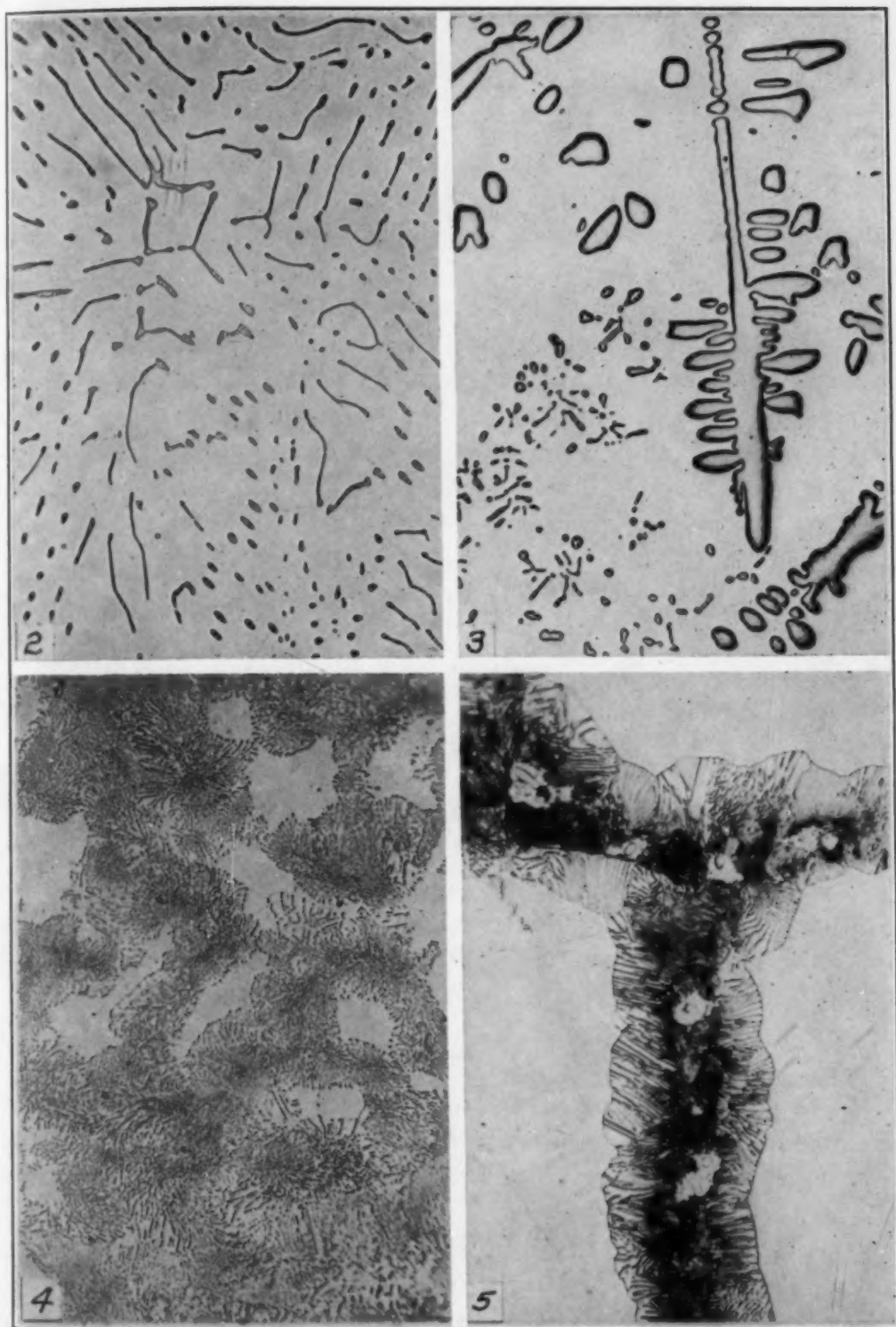


Fig. 2—Nickel 55 Per Cent, Tungsten 45 Per Cent, as Solidified from Melt. Unetched. Eutectic Structure. $\times 200$.

Fig. 3—Nickel 50 Per Cent, Tungsten 50 Per Cent, as Solidified from Melt. Unetched. Eutectic Plus Primary γ . $\times 50$.

Fig. 4—Nickel 58 Per Cent, Tungsten 42 Per Cent, Supercooled in Freezing. Electrolytic Etch, H_2O_2 + 20 Per Cent HCl. $\times 200$.

Fig. 5—Nickel 60 Per Cent, Tungsten 40 Per Cent, Supercooled in Freezing. $\times 500$.

degrees Fahr.). A 25 per cent tungsten composition showed the first indications of liquid at 1488 ± 5 degrees Cent. (2710 degrees Fahr.). It was completely melted by the time the 30 and 35 per cent tungsten alloys showed the first signs of liquid. The specimens between 30 and 40 per cent tungsten indicated the melting range to be very level and narrow between these compositions, while the maximum at 35 per cent tungsten was indicated by the 35 per cent tungsten cone which always remained standing after the 30 and 40 per cent tungsten composition had completely fused.

The solidification of the nickel-rich alloys is accompanied by marked supercooling. This effect, previously reported by Vogel, is most pronounced in the alloys of 30 to 40 per cent tungsten content. In this range, even with slow cooling the extent of supercooling may reach 100 degrees Cent. As soon as an appreciable amount of eutectic is present, however, the tendency to supercool becomes less.

A surprising structural effect is sometimes produced by the supercooling of melts in which some eutectic is present. Here the eutectic does not crystallize in the normal manner, but separates at some lower temperature in a much finer form and also in a greater quantity than should be present in that composition (Figs. 4 and 5). Such structures obviously result from conditions of nonequilibrium between liquid and solid and may be largely obliterated by heating for several hours at a temperature slightly below the solidus (Fig. 6).

The nickel-rich (α) solid solution in the recrystallized state exhibits a structure very similar to that of pure nickel (Fig. 7). As solidified from the melt the solid solution forms large polyhedral grains with occasional twins. The characteristic structure of alpha does not appear to change with the tungsten concentration. An etchant made up of 2 parts of 20 per cent HCl plus 1 part of H_2O_2 and used electrolytically with a low current density successfully develops the structure of these alloys. An equally good but less contrasty etch may be secured with a 10 per cent solution of chromic acid used electrolytically.

From the eutectic temperature down to that of the peritectoid the maximum solubility of tungsten in alpha decreases by about 2 per cent. The greatest fall occurs between the eutectic temperature and 1300 degrees Cent. (2370 degrees Fahr.) as may be seen by a comparison of Figs. 8, 9, 10. In these structures the (γ) phase has been darkened by etching in alkaline $K_3Fe(CN)_6$.

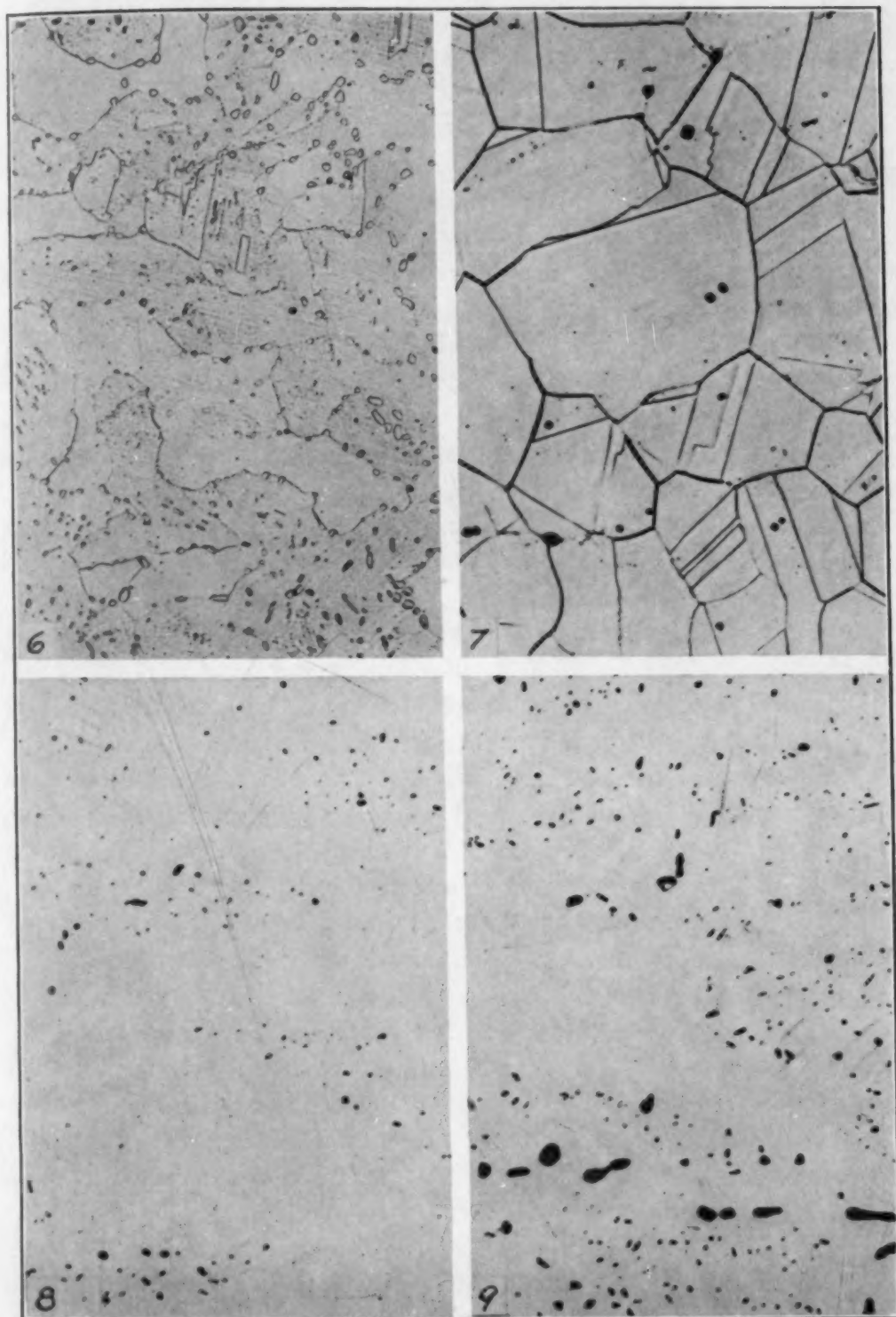


Fig. 6—Nickel 58 Per Cent, Tungsten 42 Per Cent. Part of Specimen Shown in Fig. 4 After Heating 150 Hours at 1000 to 1400 Degrees Cent. $\times 200$.

Fig. 7—Nickel 63 Per Cent, Tungsten 37 Per Cent. Rolled and Heated 80 Hours at 1000 to 1200 Degrees Cent. $\times 200$.

Fig. 8—Nickel 60 Per Cent, Tungsten 40 Per Cent. Rolled and Heated 100 Hours at 1400 Degrees Cent. Residual γ Darkened by Etch in Alkaline $K_3Fe(CN)_6$. $\times 200$.

Fig. 9—Part of Specimen Shown in Fig. 8 After Additional Heating for 100 Hours at 1300 Degrees Cent. $\times 200$.

At 970 ± 10 degrees Cent. (1780 degrees Fahr.) a peritectoid reaction occurs between the saturated alpha solid solution and the gamma phase, yielding as a product the beta phase, of approximately 43 per cent tungsten content. In alloys with less than 37 per cent tungsten content beta precipitates in a very finely divided form directly from the alpha solid solution producing the structure of Fig. 11 which darkens rapidly upon etching. It is only at temperatures just below that of the peritectoid that discrete particles of beta are produced and then only by a lengthy treatment. Prolonged heating at such temperatures results in a very little coalescence of the precipitate except at the alpha grain boundaries, as illustrated by Fig. 12.

The major decrease in the solubility of tungsten in the alpha solid solution occurs between the peritectoid temperature and 900 degrees Cent. (1650 degrees Fahr.), a smaller change in solubility being indicated between 900 and 600 degrees Cent. A solid solution of 30 per cent tungsten showed no evidence of a second phase after treating at a series of temperatures from 900 to 600 degrees Cent. (1650 to 1110 degrees Fahr.), the time at temperature being cumulative and amounting to at least 50 hours at each temperature.

The beta phase in massive form is observed in alloys of more than 37 per cent tungsten. In Fig. 13 is shown the structure of a 38 per cent tungsten alloy formed originally by sintering the pressed powder mixture at 1450 degrees Cent. (2640 degrees Fahr.) for 2 hours. This was subsequently heated for 50 hours at each of a series of temperatures decreasing by intervals of 50 degrees Cent. from 950 to 800 degrees Cent. (1740 to 1470 degrees Fahr.). A heavy precipitation of beta has taken place in the alpha matrix which is darkened by the etch. Massive beta has formed in spots originally marked by small particles of tungsten which were undissolved by the sintering treatments. The grain size of beta in massive form is considerably smaller than that of the pre-existing alpha and to outline the beta grain boundaries requires a relatively deep etch. These characteristics may be observed from a comparison of Fig. 7 with Figs. 16, 17 and 18.

The massive beta at the alpha grain boundary in a 38 per cent tungsten alloy shown in Fig. 14 is seen to be continuous with the more finely divided form of beta which separated from alpha at temperatures below 950 degrees Cent. (1740 degrees Fahr.). This beta formation occurs in spots previously occupied by gamma which

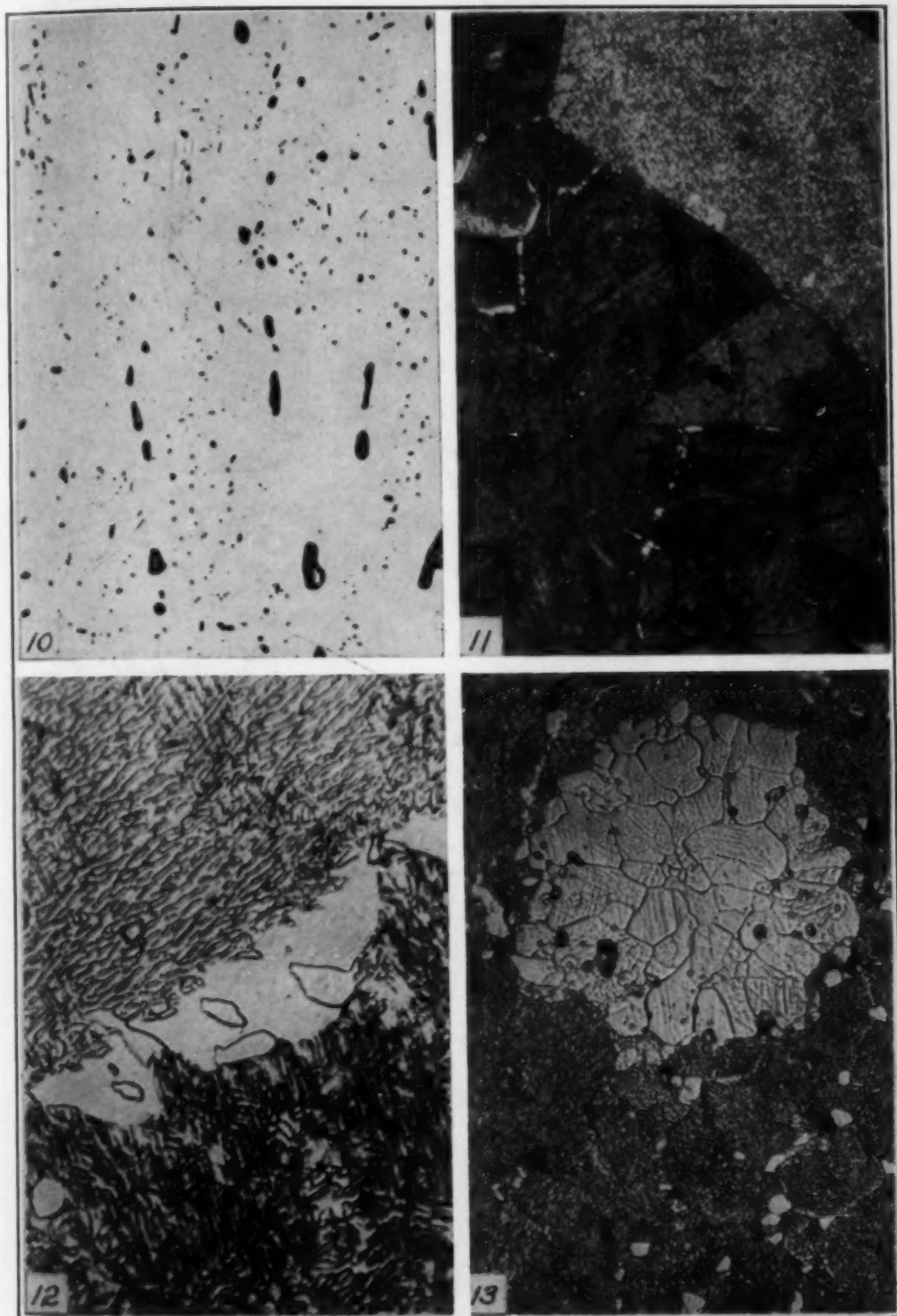


Fig. 10—Part of Specimen Shown in Fig. 9 After Additional Heating for 100 Hours Each at 1200, 1100 and 1000 Degrees Cent. $\times 200$.

Fig. 11—Nickel 63 Per Cent, Tungsten 37 Per Cent. Rolled, Heated 80 Hours at 1200 Degrees Cent. and 125 Hours at 900 Degrees Cent. $\times 200$.

Fig. 12—Nickel 62 Per Cent, Tungsten 38 Per Cent. Heated 50 Hours at 1400 Degrees Cent., 50 Hours at 700 Degrees Cent. and 100 Hours at 950 Degrees Cent. $\times 1500$.

Fig. 13—Nickel 62 Per Cent, Tungsten 38 Per Cent. Alloy Formed by Sintering at 1450 Degrees Cent. Rolled at 1200 to 1300 Degrees Cent. Heated 50 Hours Successively at 950, 900, 850 and 800 Degrees Cent. β in Both Massive (Light) and Finely Divided Form. $\times 500$.

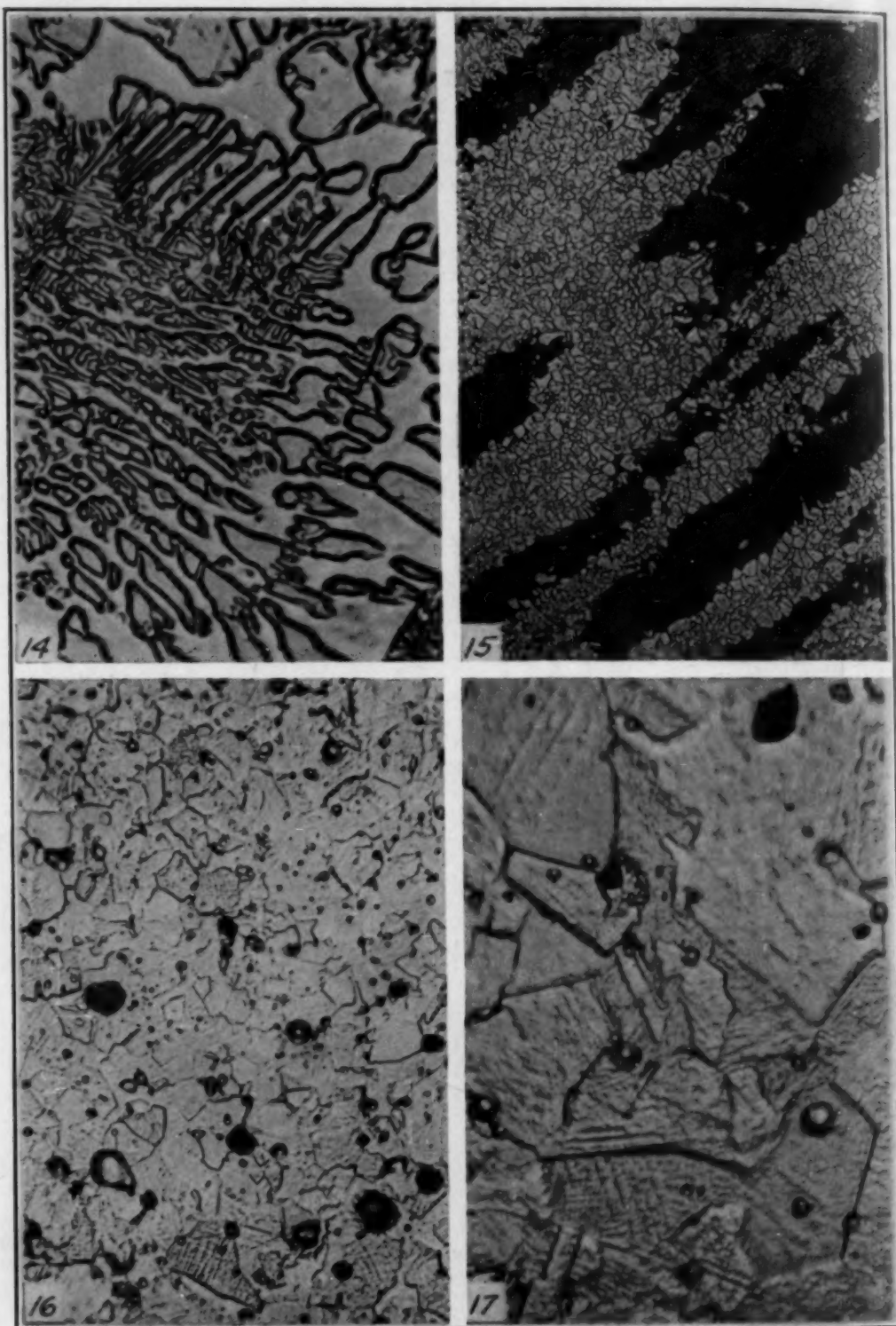


Fig. 14—Nickel 62 Per Cent, Tungsten 38 Per Cent. Alloy Formed by Melting. Rolled. Heated 40 Hours at 1000 to 1050 Degrees Cent. Heated for 25 Hours Successively at 950, 900, 850, 800, 700, 900 and 950 Degrees Cent. Massive β Agglomerating at a Grain Boundaries. $\times 1500$.

Fig. 15—Nickel 59 Per Cent, Tungsten 41 Per Cent. Rolled at 1200 to 1300 Degrees Cent. Heated for 50 Hours Successively at 950, 900, 850 and 800 Degrees Cent. $\times 200$.

Fig. 16—Nickel 57 Per Cent, Tungsten 43 Per Cent. Sintered 2 Hours at 1450 Degrees Cent. Heated Successively 100 Hours at 950 Degrees Cent., 50 Hours at 1000 Degrees Cent. and 50 Hours at 950 Degrees Cent. $\times 500$.

Fig. 17—Same Specimen as in Fig. 16. $\times 1500$.

had been precipitated from alpha by prolonged heating at 1050 to 1000 degrees Cent. (1920 to 1830 degrees Fahr.

In Fig. 15 is shown the "banded" structure of a 41 per cent tungsten alloy made by melting, which after hot rolling was heated for 200 hours between 950 and 800 degrees Cent. (1740 and 1470 degrees Fahr.). Here the nonhomogeneous nature of the original structure is indicated by the localized form of the lighter massive beta, which has developed thus in the tungsten-rich portions of the section, and discloses the directional effect of the rolling operations. The increase in amount of beta accompanying the higher tungsten content is evident from this structure when directly compared with Fig. 12. After thoroughly mixing and ballmilling the powders, a sintering treatment at 1450 degrees Cent. (2640 degrees Fahr.) followed by prolonged heating just below the peritectoid temperature results in the formation of a structure composed almost entirely of beta at the composition of 43 per cent tungsten. Such structure, approximating a single phase, is seen in Figs. 16 and 17 in which the presence of voids suggests the semiporous texture of such sintered compositions. With subsequent heating at 1100 degrees Cent. (2010 degrees Fahr.) the two phases, alpha and gamma, appear as in Fig. 18, formed by the decomposition of beta.

Progress of the peritectoid reaction over a period of time is illustrated by the structures in Figs. 19, 20 and 21. These represent an alloy of eutectic composition which after melting and hot rolling was heated at 950 degrees Cent. (1740 degrees Fahr.) for periods of 2, 5 and 25 hours respectively. The reaction rims of beta forming around the elongated gamma of the eutectoid structure are distinctly visible in Figs. 19 and 20. In such a structure the reaction approaches completion only after some 500 hours heating near the peritectoid temperature, and the resulting structure in a 45 tungsten alloy is seen in Fig. 22 wherein the slight excess of gamma may be observed. Progress of this reaction in an unworked alloy of like composition almost ceases after 50 hours at 950 degrees Cent. (1740 degrees Fahr.) due very likely to the relatively large distances separating the reacting phases as illustrated by the structure in Fig. 23.

The sintered structure in alloys of higher tungsten content is illustrated in Fig. 24, in which a relief polish reveals the islands of alpha scattered through a gamma matrix. Upon heating this sintered alloy to 1500 degrees Cent. (2730 degrees Fahr.), the formation of

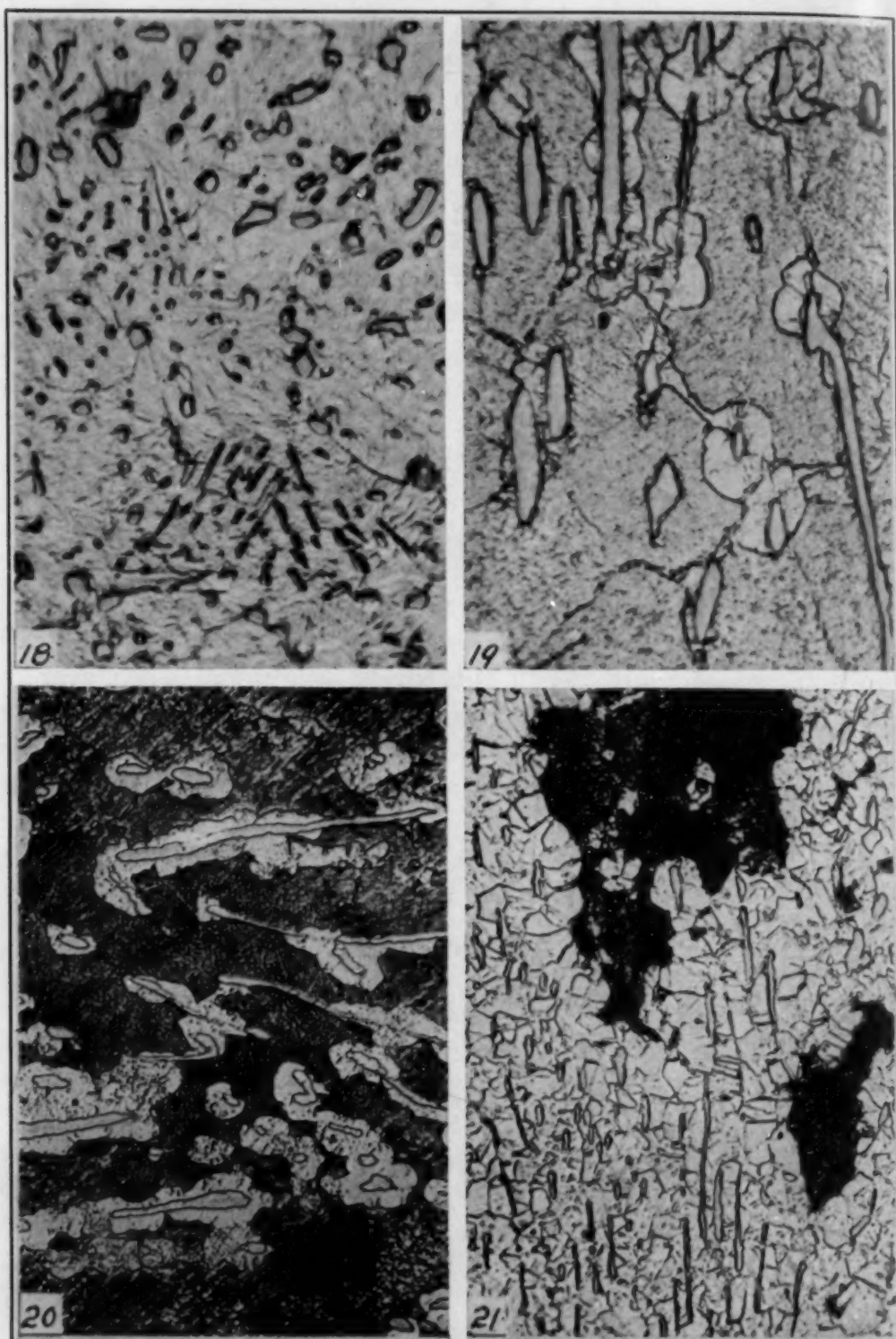


Fig. 18—Same Specimen as in Fig. 17 After Heating 60 Hours at 1100 Degrees Cent. $\alpha + \gamma$. $\times 1500$.

Fig. 19—Nickel 55 Per Cent, Tungsten 45 Per Cent. Alloy Formed by Melting. Rolled 1200 to 1300 Degrees Cent. Heated 2 Hours at 950 Degrees Cent. Reaction Rims Surrounding γ . $\times 1500$.

Fig. 20—Same Specimen Shown in Fig. 19 After 5 Hours at 950 Degrees Cent. $\times 500$.

Fig. 21—Same Specimen Shown in Figs. 19 and 20 After 25 Hours at 950 Degrees Cent. γ Almost Completely Replaced by β . $\times 500$.

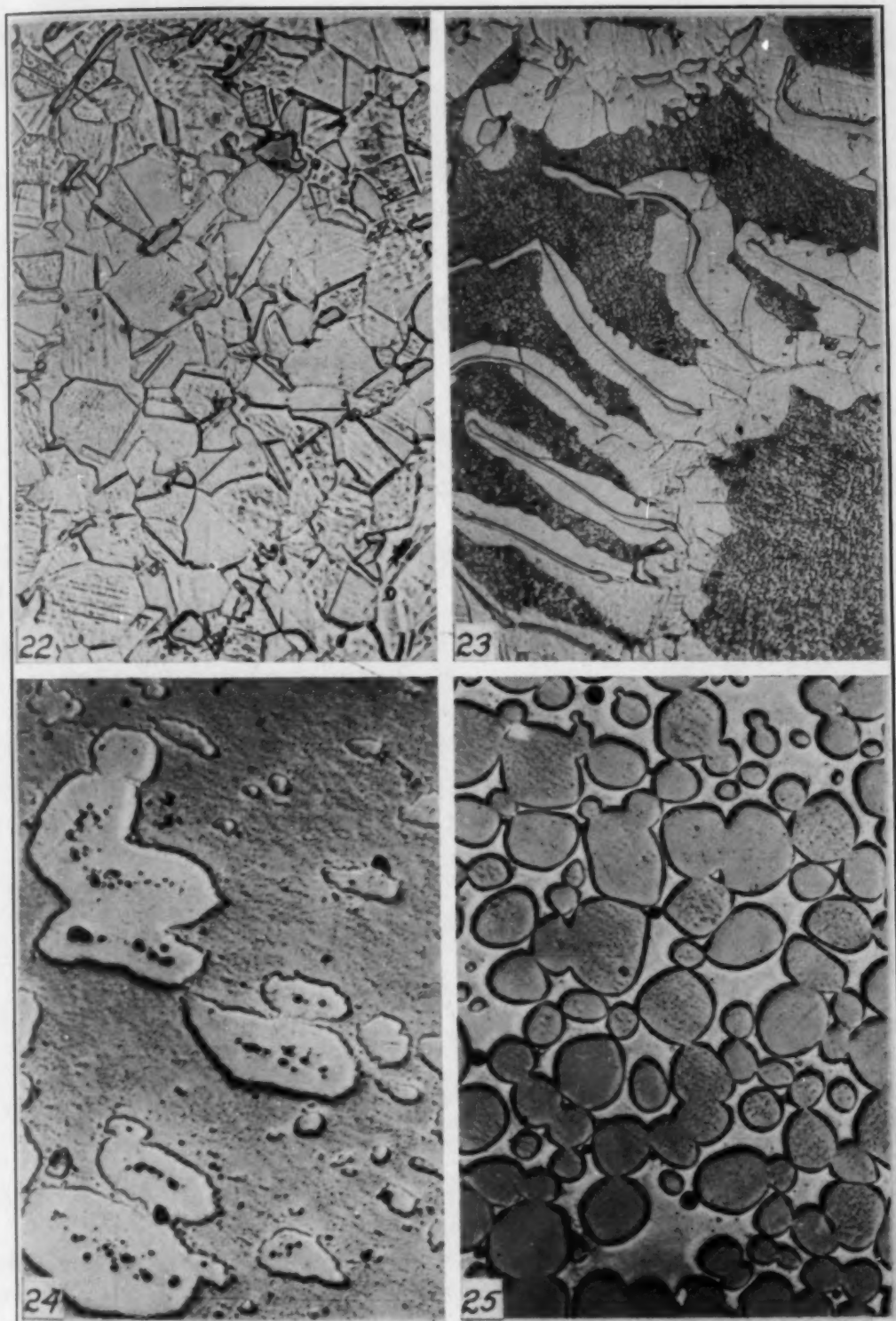


Fig. 22—Nickel 55 Per Cent, Tungsten 45 Per Cent. Alloy Formed by Melting. Rolled 1200 to 1300 Degrees Cent. Heated Successively 100 Hours at 900, 950, 1000, 975 and 950 Degrees Cent. β + Slight Excess of γ . $\times 500$.

Fig. 23—Nickel 55 Per Cent, Tungsten 45 Per Cent, Unworked Alloy. Heated 100 Hours at 950 Degrees Cent. $\times 500$.

Fig. 24—Nickel 10 Per Cent, Tungsten 90 Per Cent, Metal Powders Mixed, Pressed and Sintered at 1450 Degrees Cent. Unetched. α (Light Islands) + γ . $\times 200$.

Fig. 25—Same Alloy as Shown in Fig. 24. Sintered 1 Hour at 1500 Degrees Cent. Unetched. γ (Darker Islands) + α Matrix. $\times 200$.

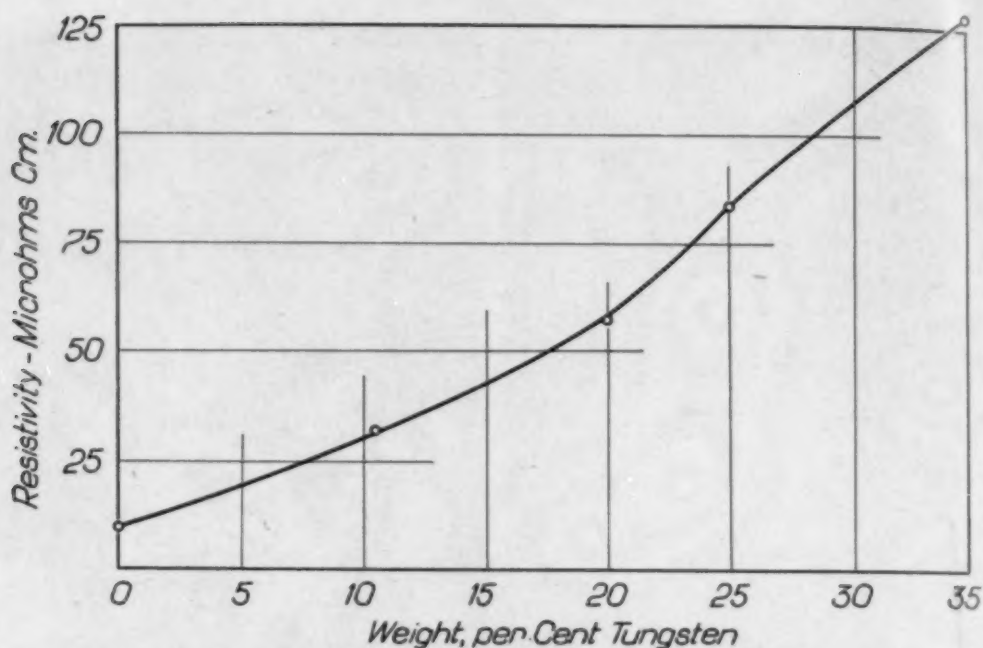


Fig. 26—Electrical Resistivity of Nickel Rich Solid Solutions.

the liquid phase changes the pattern of the structure to that seen in Fig. 25. Here the rounded masses of solid gamma were completely surrounded by the nickel-rich liquid phase which appears now in the unetched section as the lighter alpha matrix. This structure is typical of tungsten-rich and molybdenum-rich compositions in several other binary systems after similar treatment, and has recently been discussed in detail by Price, Smithells and Williams (5).

ELECTRICAL RESISTIVITY

Wires 0.025 inch in diameter were obtained by rolling, swaging and drawing alloys of 10, 20, 25 and 35 per cent tungsten. Elec-

Table I
Electrical Resistivity of Nickel-Tungsten Alloy Wires

Treatment	Microhm.-cm.			
	10% Tungsten	20% Tungsten	25% Tungsten	35% Tungsten
1200 Degrees Cent.— 5 hours..... (2190 Degrees Fahr.)	32.1	57.3	83.3	127.1
1200 Degrees Cent.— 5 hours..... (2190 Degrees Fahr.)
950 Degrees Cent.—100 hours..... (1740 Degrees Fahr.)	57.7	83.6	127.1
900 Degrees Cent.—100 hours..... (1650 Degrees Fahr.)	58.0	83.6	104.9
800 Degrees Cent.—100 hours..... (1470 Degrees Fahr.)	57.5	83.4	82.0
700 Degrees Cent.—100 hours..... (1290 Degrees Fahr.)	57.3	83.4	75.0

trical resistivity determinations were made upon wire specimens after homogenizing at 1200 degrees Cent. (2190 degrees Fahr.) and cooling rapidly to room temperature, both heating and cooling being carried out in a hydrogen atmosphere to protect the surface of the wire. The apparatus and method employed in these measurements have been fully described elsewhere (4). Resistivities of this series of solid solutions are listed in Table I and plotted against tungsten concentration in Fig. 26. To confirm the lower boundary of the alpha field as determined from the microstructure the alloy

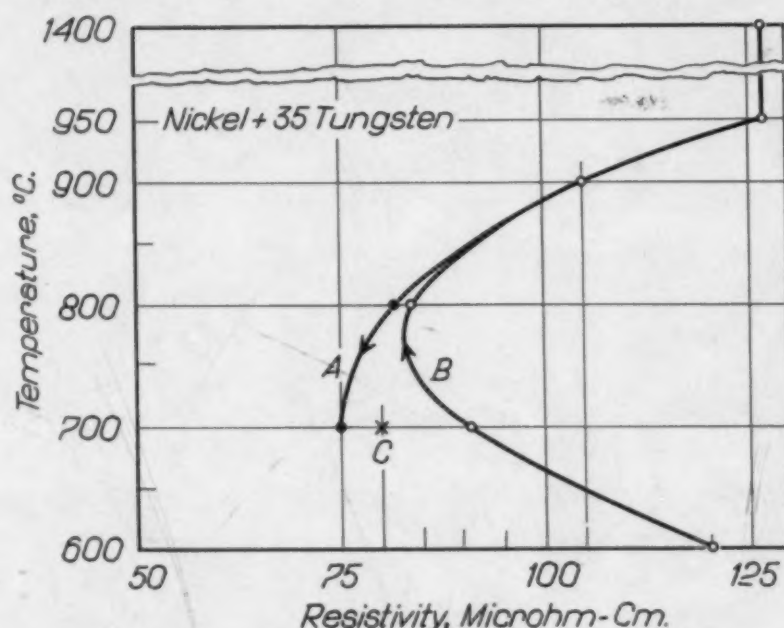


Fig. 27—Electrical Resistivity of Nickel + 35 Per Cent Tungsten Alloy. Homogenized at 1200 Degrees Cent. and Cooled Rapidly to Room Temperature.

A. Heated for 100 Hours at Each of the Indicated Series of Decreasing Temperatures. Time Cumulative.

B. Heated for 100 Hours at Each of Indicated Increasing Temperatures. Time Not Cumulative.

C. Heated for 400 Hours at 700 Degrees Cent.

wires were heated for 100 hours successively at 950, 900, 800 and 700 degrees Cent. (1740, 1650, 1470 and 1290 degrees Fahr.), resistivity measurements being made at 25 degrees Cent. following treatment at each temperature. The alloy of 35 per cent tungsten alone showed a marked decrease in resistivity after the treatment at 900 degrees Cent. (1650 degrees Fahr.), as might be expected from the solubility curve as established by observations of microstructure.

The progressive fall in resistivity of this alloy resulting from 100-hour treatments at decreasing temperatures is shown by curve A of Fig. 27. The relatively slight decrease between 800 and 700

degrees Cent. (1470 and 1290 degrees Fahr.) probably indicates a correspondingly small change in the solubility of tungsten in nickel between those temperatures. A rapid lowering of the diffusion rate with decreasing temperature would, of course, tend to produce a similar effect. The influence of this factor is illustrated by the course of curve B in Fig. 27. Here are plotted the resistivities observed after heating for 100 hours at a series of increasing temperatures.

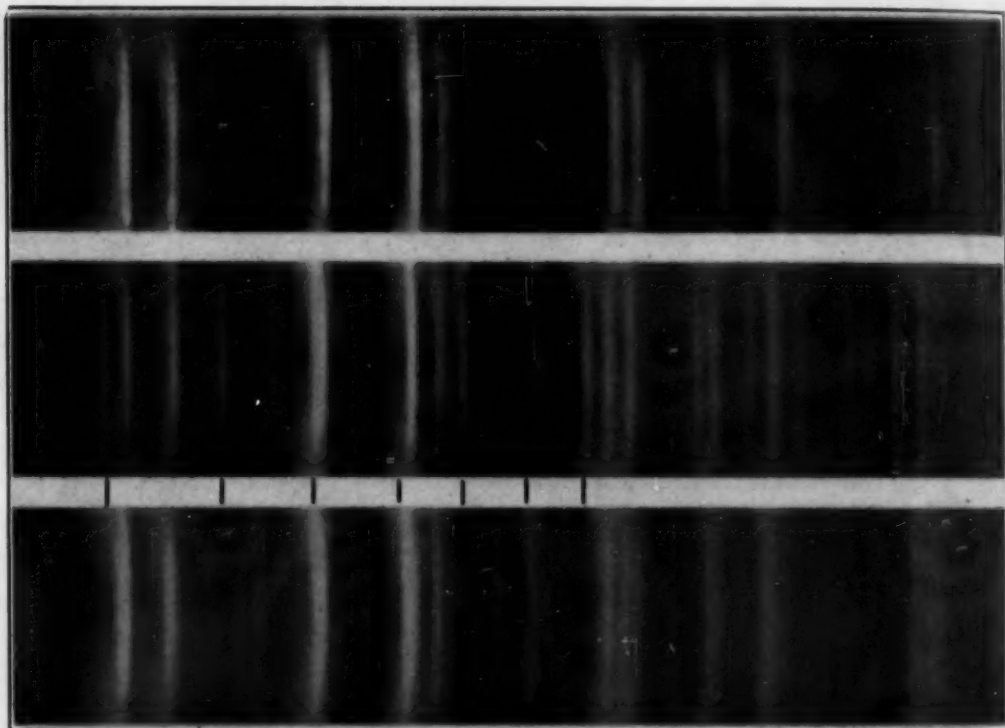


Fig. 28—Diffraction Patterns Characteristic of the Three Phases Observed in this Alloy System.

Upper—Nickel 90 Per Cent, Tungsten 10 Per Cent. Alpha Solid Solution in Sintered Alloy.

Middle—Nickel 50 Per Cent, Tungsten 50 Per Cent. Reflection from Alpha and Gamma in Alloy as Sintered at 1400 Degrees Cent., Several Reflections from Gamma Underlined.

Bottom—Nickel 57 Per Cent, Tungsten 43 Per Cent. Reflections from Beta Formed by Heating at 950 Degrees Cent. for 150 Hours, After First Sintering at 1400 Degrees Cent.

In obtaining the values for curve B separate sets of three specimens each were heated at each temperature, i.e., one set at 600 degrees Cent. (1110 degrees Fahr.), a second at 700 degrees Cent. (1290 degrees Fahr.) etc., subsequent to the homogenizing treatment at 1200 degrees Cent. (2190 degrees Fahr.). In all cases the recorded values represent the average of five readings on each of three specimens treated together. From the course of curve B it

appears that equilibrium is not attained during a period of 100 hours at a temperature as high as 800 degrees Cent. (1470 degrees Fahr.) After 400 hours at 700 degrees Cent. (1290 degrees Fahr.) the resistivity has decreased to the value indicated at point C in Fig. 27, still somewhat greater than that measured after the cumulative periods of heating at 900, 800 and 700 degrees Cent. (1650, 1470 and 1290 degrees Fahr.).

X-RAY DIFFRACTION PATTERNS

Many of the alloy specimens treated for metallographic examination were also used to obtain diffraction patterns. Reflections were observed from three phases only throughout the system. These

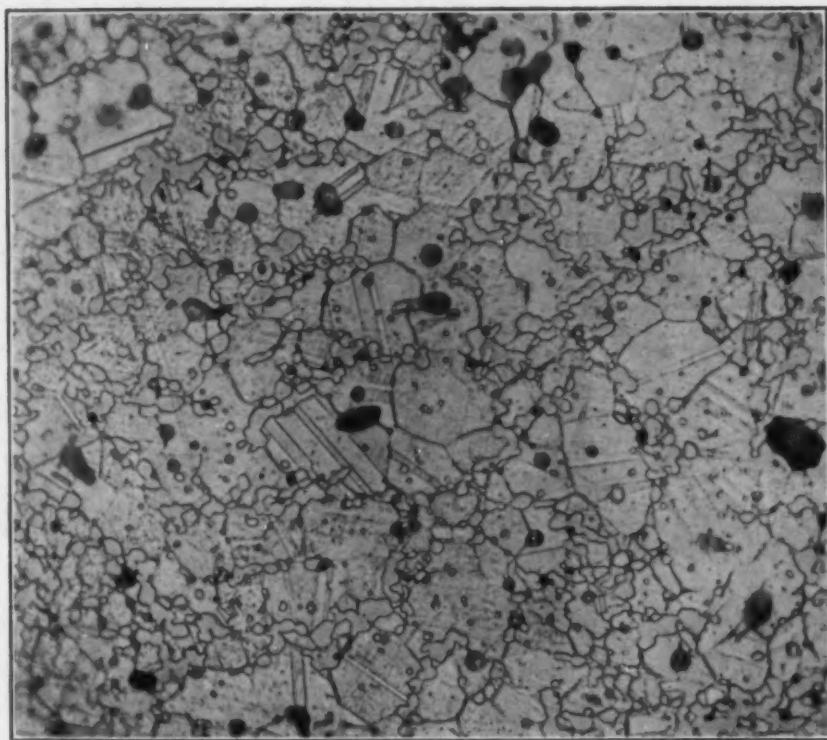


Fig. 29—Nickel 45 Per Cent, Tungsten 55 Per Cent. Sintered—1450 Degrees Cent. $\times 500$.

are shown in Fig. 28. The upper photogram is representative of the 10 per cent tungsten alloy, a face-centered cubic pattern of the alpha solid solution. When saturated with tungsten the parameter of the alpha lattice approximates 3.65 Å, exceeding by some 4 per cent the corresponding value of 3.52 Å shown by the nickel powder. An alloy of 50 per cent tungsten after sintering at 1400 degrees Cent.

(2550 degrees Fahr.) yields the characteristic reflections of both alpha and gamma, the latter being identical in position with those of tungsten within the precision limits of the methods employed in this investigation. The middle pattern of Fig. 28 represents this alloy in such a condition. Here the intense reflections from alpha may be readily identified with the corresponding lines from the same phase of lower tungsten content directly above. The displacement resulting from increased tungsten concentration is plainly visible.

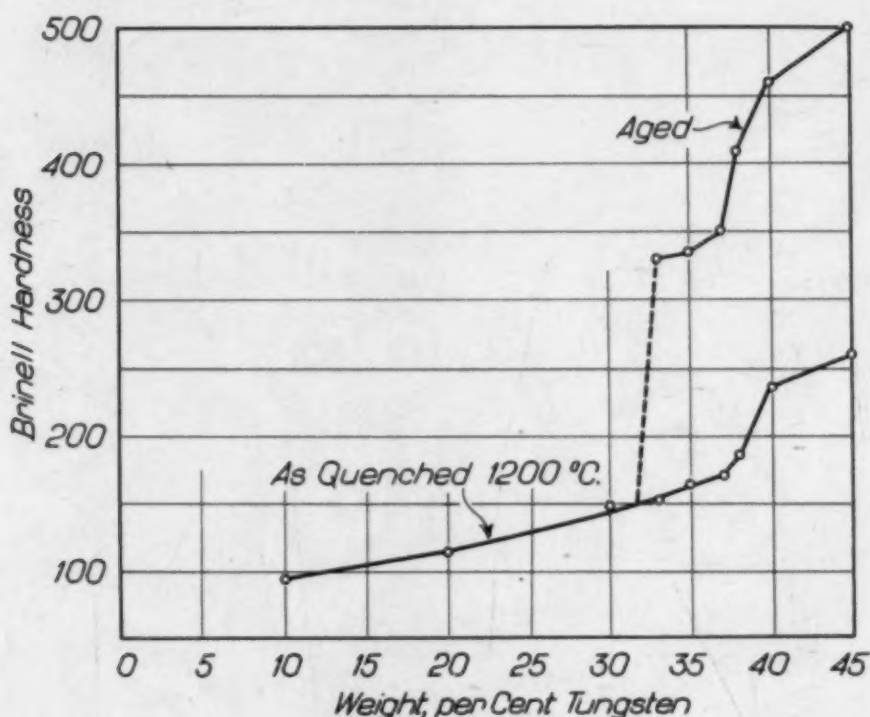


Fig. 30—Hardness vs. Tungsten Concentration in Nickel-Tungsten Alloys.

Five gamma reflections are marked immediately below this photograph, which is representative of the microstructure shown in Fig. 29.

In an alloy of this composition heated at temperatures slightly below 970 degrees Cent. (1780 degrees Fahr.) the beta phase is formed by the peritectoid reaction and gamma is replaced by beta wholly or in part depending upon the composition. The lower photograph in Fig. 28 was prepared from a 43 per cent tungsten alloy, after such a treatment, and corresponds to the microstructure shown in Figs. 16 and 17. After sintering at 1400 degrees Cent. (2550 degrees Fahr.) this composition yields a pattern identical with the middle one in Fig. 28 except for a slightly lower intensity of the gamma

reflections. As a result of the lower temperature treatment, the alloy is converted almost completely to beta, as the microstructure shows, and the reflections in the lower photogram are characteristic of the beta phase. It will be observed that the more intense reflections have shifted hardly at all from their position in the alpha pattern but are more diffuse. A new series of less intense reflections

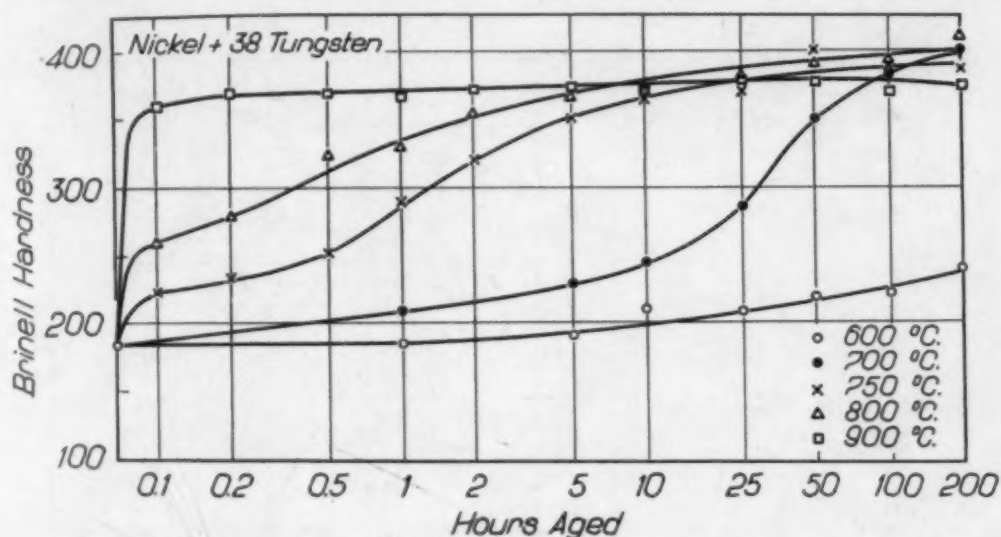


Fig. 31—Nickel 62 Per Cent, Tungsten 38 Per Cent. Alloy Rolled, Homogenized at 1200 Degrees Cent. and Aged at Temperatures Shown.

has replaced those of gamma which appear in the middle photogram. Casual analysis suggests the formation of a superlattice upon the pre-existing face-centered cubic lattice of the alpha phase. Since a new phase is observed in the microstructure, however, this terminology might be ruled out on ground of definition.

PRECIPITATION HARDENING

The shape of the solid solubility curve over the range of composition extending from 30 to 40 per cent tungsten suggests that such alloys may be hardened by a precipitation treatment.

In Fig. 30 the lower line represents the hardness of the one and two phase structures as quenched from 1200 degrees Cent. (2190 degrees Fahr.) over the composition range 10 to 45 per cent tungsten. To cover the full range of hardness in this series both B and C scales of the Rockwell system were required. These readings have been converted to Brinell numbers for plotting in the accompanying charts.

The 30 per cent tungsten alloy showed no response to aging at

any temperature between 900 and 600 degrees Cent. (1650 and 1110 degrees Fahr.), while those containing from 33 to 45 per cent tungsten began to harden slowly after some 10 hours at 600 degrees Cent. (1110 degrees Fahr.) as shown by the lower curves in Figs. 31 and 32. The rapid increase in hardening rate which accompanies a rise in aging temperature from 700 to 750 degrees Cent. (1290 to 1380 degrees Fahr.) is indicated in Fig. 33, drawn from the data obtained

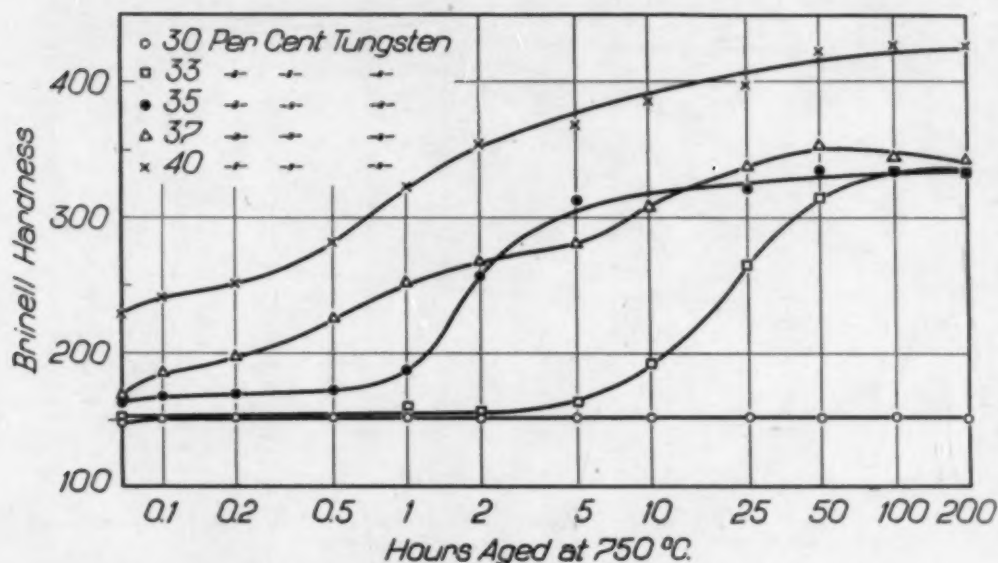


Fig. 32—Nickel-Tungsten Alloys of Various Concentrations. Rolled, Homogenized at 1200 Degrees Cent. and Aged at 750 Degrees Cent.

on the 38 per cent tungsten alloy. One unusual feature exhibited by these alloys is the persistence of the hardness throughout extended periods of aging at temperatures as high as 900 degrees Cent. (1650 degrees Fahr.). This is clearly shown by the time-hardness curves from the 38 per cent tungsten alloy in Fig. 31. It hardens the more rapidly as the aging temperature increases in a conventional manner, but the hardness values at the end of 200 hours at 700 to 900 degrees Cent. (1290 to 1650 degrees Fahr.) are nearly identical and there is only slight indication of over-aging at 900 degrees Cent.

A second example of uncommon performance is shown in Fig. 34. Here the 33 per cent tungsten alloy is seen to harden more rapidly at 750 degrees Cent. (1380 degrees Fahr.) than at 800 degrees Cent. (1470 degrees Fahr.), although over a period of 500 hours the hardness resulting from the lower temperature treatment is distinctly the greater of the two, thus conforming in part of the relationship usually observed.

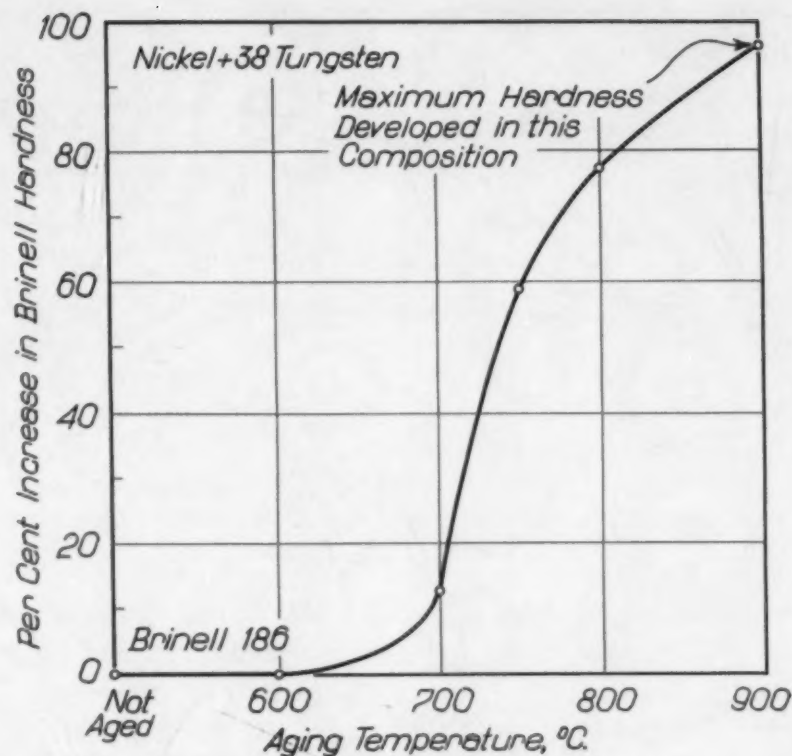


Fig. 33—Nickel 62 Per Cent, Tungsten 38 Per Cent. Hardness After 1 Hour Aging at Various Temperatures.

An intermediate composition of 35 per cent tungsten, when aged at a series of temperatures extending from 900 to 750 degrees Cent. (1650 to 1380 degrees Fahr.) shows characteristics common to both the 33 and the 38 per cent tungsten alloys. Such a set of time-

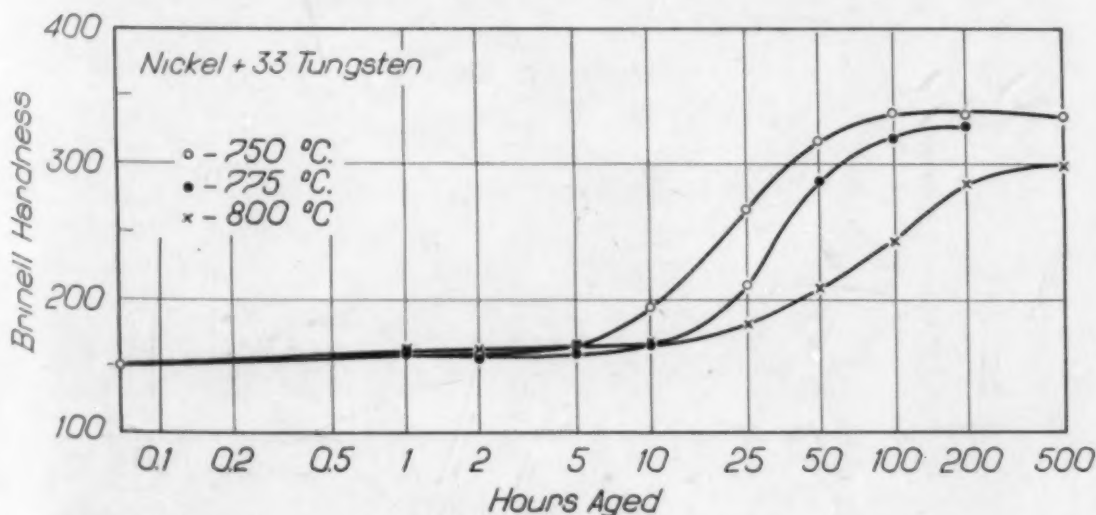


Fig. 34—Nickel 67 Per Cent, Tungsten 33 Per Cent, Rolled, Homogenized at 1200 Degrees Cent., Aged at 750, 775 and 800 Degrees Cent.

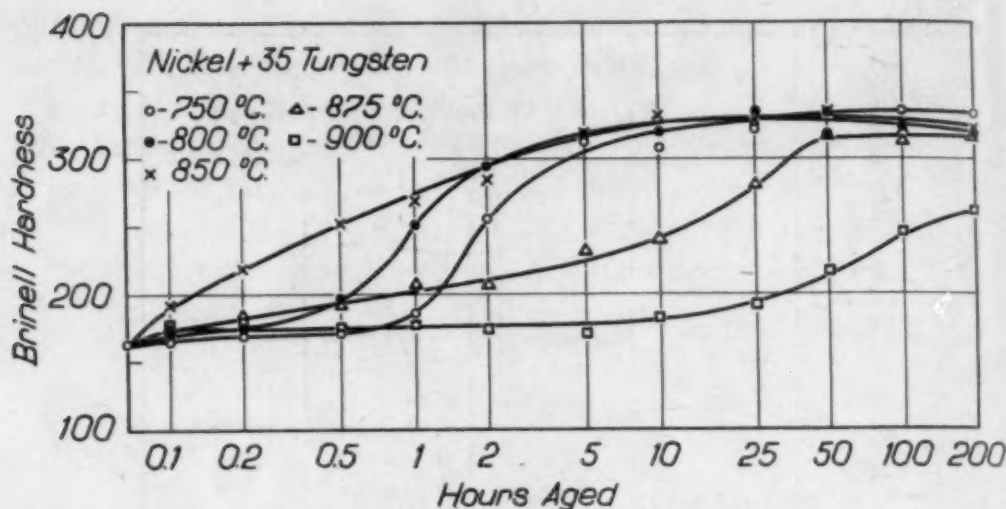


Fig. 35—Nickel 65 Per Cent, Tungsten 35 Per Cent. Rolled, Homogenized at 1200 Degrees Cent. and Aged at Temperatures Shown.

hardness curves appears in Fig. 35. Hardening here sets in most rapidly at 850 degrees Cent. (1560 degrees Fahr.) and most slowly at 900 degrees Cent. (1650 degrees Fahr.). The curves for 800 and 750 degrees Cent. (1470 and 1380 degrees Fahr.) both rise rapidly after the first half hour and pass above the 875 degrees Cent. (1605 degrees Fahr.) curve. At the end of 50 hours the hardness values resulting from aging treatment at the four temperatures below 900 degrees Cent. (1650 degrees Fahr.) are approximately identical

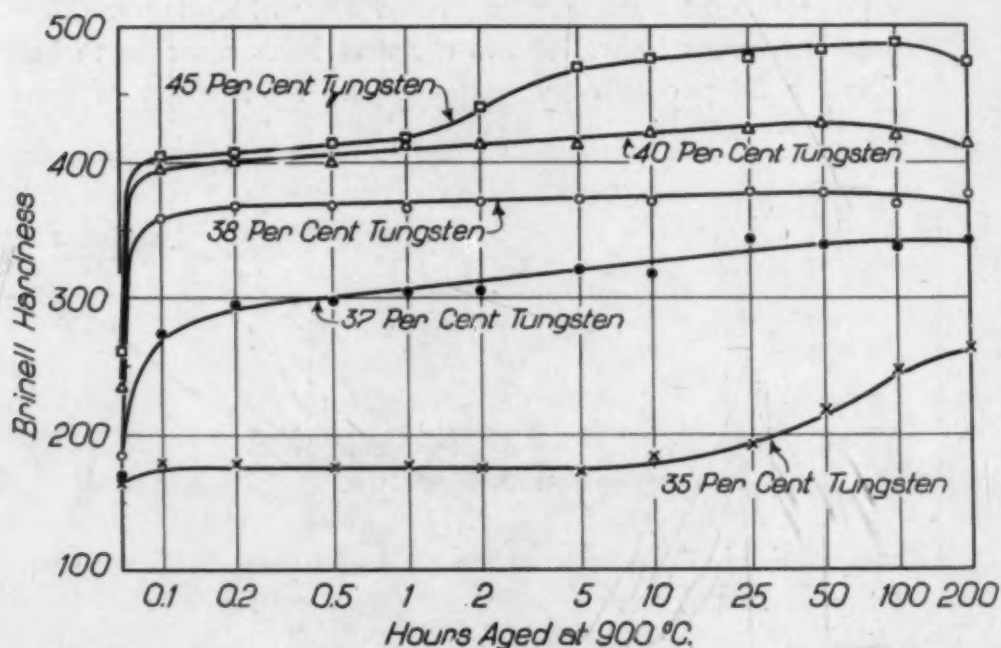


Fig. 36—Nickel Alloys of Various Tungsten Concentrations. Rolled, Homogenized at 1200 Degrees Cent. Aged at 900 Degrees Cent.

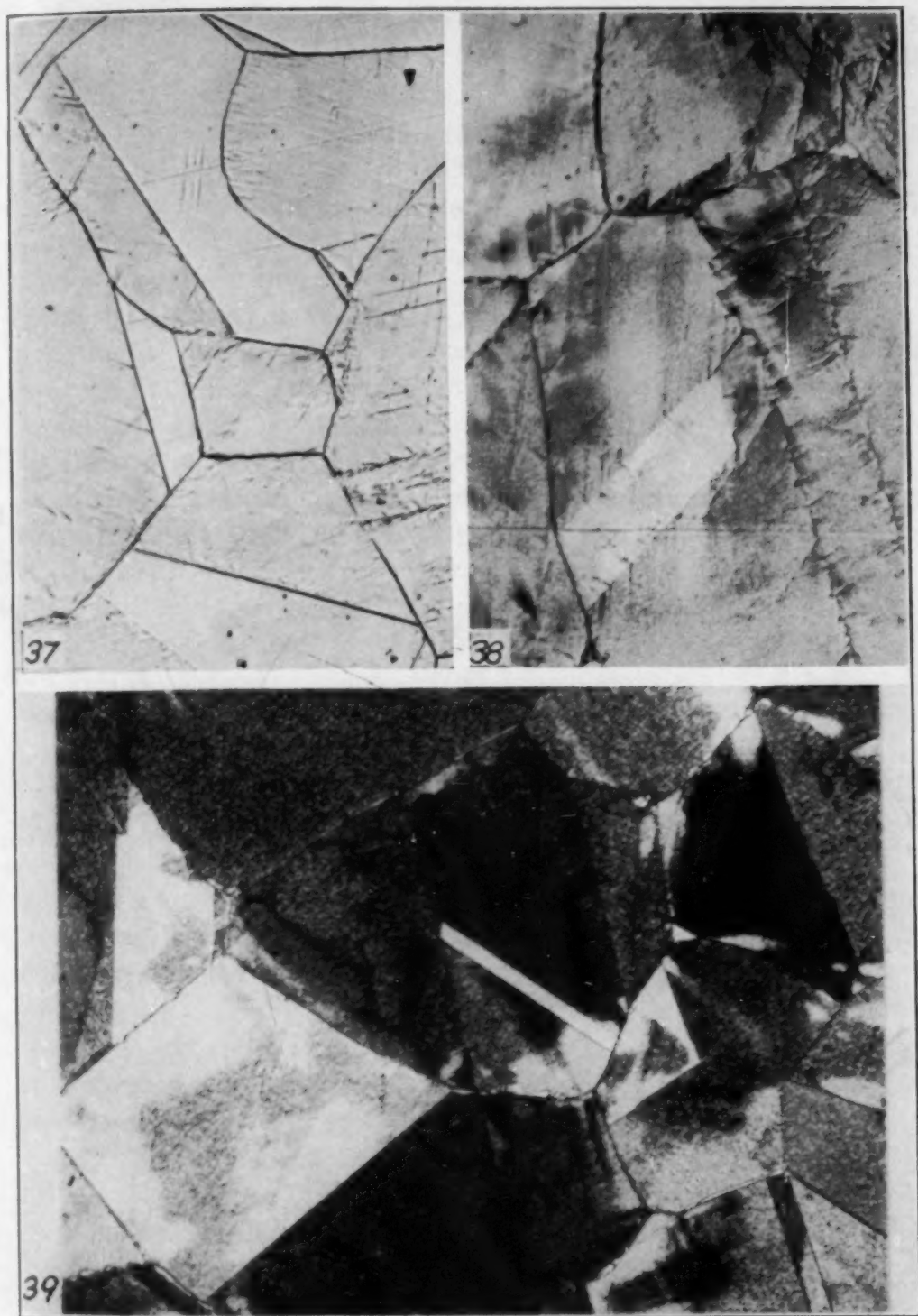


Fig. 37—Nickel 65 Per Cent, Tungsten 35 Per Cent. Rolled, Homogenized at 1200 Degrees Cent. for 5 Hours, Quenched and Aged for 200 Hours at 750 Degrees Cent. $\times 200$.

Fig. 38—Same Alloy as Shown in Fig. 37. Aged 500 Hours at 800 Degrees Cent. $\times 200$.

Fig. 39—Same Alloy as Shown in Figs. 37 and 38. Aged 200 Hours at 900 Degrees Cent. $\times 200$.

although here again the highest hardness seems to accompany the lowest aging temperature.

The rapid response of the higher tungsten compositions to aging at elevated temperatures is clearly illustrated by the time-hardness curves in Fig. 36. Of this series of alloys, aged at 900 degrees Cent. (1650 degrees Fahr.), the 35 per cent tungsten specimens begin to harden with more delay and at a lower rate than those of higher tungsten content. The increase in hardening rate shown by the 45 per cent tungsten alloy between the 1- and 5-hour periods may possibly be associated in some manner with the formation of the massive beta areas seen in the microstructures of Figs. 19 and 20.

It may well be that the unusual hardening characteristics of these alloys arise from the fact that the quantity of beta increases very rapidly over a narrow range of tungsten concentration, i.e., between 30 and 43 per cent tungsten by weight the amount of this phase increases from 0 to 100 per cent in the structure. Furthermore, the mechanism of precipitation hardening may be modified by the peritectoid reaction which involves both alpha and beta.

The microstructures of the 35 per cent tungsten alloy after 200 hours at 750, 800 and 900 degrees Cent. (1380, 1470 and 1650 degrees Fahr.) are shown in Figs. 37, 38 and 39. As aged to maximum hardness at 750 degrees Cent. (1380 degrees Fahr.) the alpha solid solution is marked by a Widmanstätten pattern plainly seen in Fig. 37. A slight thickening of the grain boundaries is likewise noticeable but etching produces no general darkening. After aging to approximately the same hardness at 800 degrees Cent. (1470 degrees Fahr.) for 500 hours, the second phase is visible at the grain boundaries in larger amounts while the alpha grains begin to show local darkening after the etch as illustrated in Fig. 38. Aging at 900 degrees Cent. (1650 degrees Fahr.) for 200 hours produces the structure seen in Fig. 39 which is characterized by further increase of grain boundary precipitate, and more especially by the rapidity of etch which darkens the matrix extensively. The structures shown in Figs. 37 and 38 suggest a surface broken by lines and areas of varying elevation such as might result from plastic deformation.

ACKNOWLEDGMENT

The authors are much indebted to Mr. A. B. Gladding for the many X-ray photograms prepared during the course of this investigation.

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2. R. Vogel, "Über Wolfram-Nickellegierungen," *Z. anorg. allg. Chem.* Vol. 116, 1921, p. 231.
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4. W. P. Sykes, "The Cobalt-Tungsten System," *TRANSACTIONS, American Society for Steel Treating*, Vol. 21, 1933, p. 416.
5. G. H. S. Price, C. J. Smithells and S. V. Williams, "Sintered Alloys," Part I. Copper-Nickel-Tungsten Alloys Sintered with a Liquid Phase Present, *Journal, Institute of Metals*, Vol. 62, No. 1, 1938, p. 239.

DISCUSSION

Written Discussion: By Dr. R. H. Harrington, research metallurgist, General Electric Co., Schenectady, N. Y.

Equilibrium systems, involving molybdenum, tungsten or cobalt are usually characterized by slow rates of diffusion resulting in the most difficult research conditions in the field of studies of equilibria. As usual with work associated with the name of Sykes, one may approach the review of this report with utmost confidence in the presented results and a real pleasure in the reading thereof.

From the evidence offered there can be no doubt that the peritectoid reaction as reported truly exists in the place of the eutectoid reaction previously reported by R. Vogel as long ago as 1921. That there should otherwise be such close agreement between these two published diagrams should be mutually satisfying to both the authors of this report and to Dr. Vogel.

One or two minor points might be discussed:

1. Photomicrograph No. 7 is described as representative of the α -nickel-rich solid solution. The appearance of the grain boundaries suggests the possible presence there of a second phase. The half-tone prints do not allow for suitable magnification of the published prints. The writer will be satisfied with a definite statement from the authors regarding this point.

2. As the authors state, the structures shown in Figs. 37 and 38 suggest the occurrence of plastic deformation. Such plastic deformation very possibly could result from water quenching or even rapid air cooling from the temperatures of the heat treatments. In all cases, what was the rate of cooling from the noted heat treatment temperatures?

3. An instance of typographical omission is the absence of the magnification for Fig. 39. It seems not unreasonable to assume $\times 200$, the same magnification as for Figs. 37 and 38.

There is a growing interest in the ternary system of Fe-Ni-W. It would be an excellent service if the authors would continue their work to readjust the results reported by Winkler and Vogel to agree with this latest information of the Ni-W binary system.

Written Discussion: By R. H. Schaefer, research laboratory, The International Nickel Co., Inc., Bayonne, N. J.

In determining the constitutional diagrams of metal systems, the use of

more than one type of measurement is usually necessary. The authors of the present interesting and systematic contribution to the literature on age hardening alloys are to be commended for the variety and choice of tools which they employed to study this particular system. Some of the discrepancies between this and earlier work undoubtedly may be attributed to the reliance placed by previous investigators on inadequate data, and possibly from failure to fully appreciate the sluggishness with which these binary alloys approach their equilibrium conditions. The authors mention that they failed to find any evidence of the compound Ni_3W as reported by R. Vogel. It might be pointed out that the same conclusion was reached by Becker and Ebert¹ who studied an alloy of this composition by means of X-ray diffraction.

The curve shown in Fig. 26 indicates that with increasing tungsten content up to 20 per cent the rate of increase in resistivity also increases. This is in opposition to what has been previously found for numerous nickel-rich solid solutions. We believe it is due to the slightly high value shown for pure nickel, which may be the result of the presence of small amounts of impurities which exert less effect in the tungsten-containing alloys. Substituting data which we have collected for alloys containing less than 5 per cent of tungsten and using that of the authors at higher tungsten levels, the relationship between the amount of solute and the resistivity becomes linear up to 20 per cent tungsten. The break in the resistivity curve between 20 and 25 per cent tungsten is the result of the magnetic transformation of nickel which is lowered to room temperature by this amount of alloying addition.

The age hardening curves which result when beta (which incidentally forms quite close to the composition corresponding to Ni_3W) is rejected from supersaturated alpha solid solution are particularly interesting. As is mentioned by the authors, when an alloy containing 33 per cent tungsten was aged at 750 to 800 degrees Cent. (1380 to 1470 degrees Fahr.) hardening occurred more rapidly at the lower temperatures. Similarly in a 35 per cent tungsten alloy this occurred between 850 and 900 degrees Cent. (1560 and 1650 degrees Fahr.). This somewhat unusual performance was not encountered in a 38 per cent tungsten when aged at temperatures of from 600 to 900 degrees Cent. (1110 to 1650 degrees Fahr.). If Fig. 1 is examined it will be seen that this uncommon behaviour only occurred at low degrees of supersaturation, and is quite similar to results reported by Wise and Easch² for age-hardenable nickel bronzes for slightly supersaturated alloys.

The excellent photomicrographs shown in Figs. 37 to 39 indicate that the mechanism of precipitation in a 35 per cent tungsten alloy is at least partly dependent upon the aging temperature used. Have the authors considered the possibility that this precipitation may involve the formation of transition lattices such as are suggested by Mehl and Jetter³ in a paper on the mechanism of precipitation to be presented later at this convention?

¹Becker and Ebert, *Z. Physik*, Vol. 16, 1923, p. 168.

²E. M. Wise and J. T. Easch, "Strength and Aging Characteristics of the Nickel Bronzes", American Institute of Mining and Metallurgical Engineers, Technical Publication 523, 1934.

³R. F. Mehl and L. K. Jetter, "The Mechanism of Precipitation from Solid Solution. The Theory of Age-Hardening", Age Hardening of Metals, American Society for Metals, 1939, p. 342.

Oral Discussion

L. H. DEWALD:⁴ We have done some work at Fansteel on this particular system, and I would be interested in knowing just what difficulties you had with rolling and working these alloys in the high percentages of tungsten, like 48 per cent. I have had great difficulty in working percentages as low as thirty.

Authors' Reply

The authors wish to thank Dr. R. H. Harrington and Mr. R. H. Schaefer for their interesting discussions.

In reply to Dr. Harrington, we would say that the grain boundary appearance in Fig. 7 is caused by a heavy etch; and that the alloys prior to age hardening were all water quenched from the solution temperature.

As to the rolling characteristics of the nickel-tungsten alloys: up to 25 per cent tungsten we did not have difficulty in rolling them, but from 30 to 40 per cent tungsten we encountered the greatest difficulty. As soon as a little excess tungsten is present, the rolling quality seems to improve again. It was just in that range between 30 and 40 per cent tungsten that we had difficulty in rolling the alloys. Rolling temperatures for the higher tungsten concentrations were between 1200 and 1300 degrees Cent. (2190 and 2370 degrees Fahr.); up to 20 per cent tungsten they could be rolled at 900 to 1000 degrees Cent. (1650 to 1830 degrees Fahr.).

⁴Metallurgical engineer, Fansteel Metallurgical Corp., North Chicago, Ill.

TESTING STEEL WITH THE CUTTING FLAME

BY JAMES R. CADY

Abstract

A new method of treating steel samples for metallographic examination and testing is proposed. A properly shaped sample of the steel to be tested is cut along one edge with the oxyacetylene cutting flame. The conditions under which the cut is made are controlled to exaggerate the effect of the flame upon the steel and at the same time to give reproducible cutting conditions. A metallographic sample is sawed from this flame cut sample and polished on a plane normal to the flame cut surface. The microstructure of the flame affected edge as well as the normal structure of the metal is, in this manner, made available for examination. Microstructures produced by flame testing are similar to structures obtained by heating the steel and quenching at various temperatures from the critical range to the melting point. In addition to these quench structures, a hypereutectoid structure resulting from carburizing the steel along the cut surface is formed. This carburized zone is similar in appearance to the hypereutectoid zone of case carburized steel samples. The presence of certain alloys in the steel is revealed by an austenitic zone occurring on the flame cut surface and replacing the usual carburized zone.

A STUDY of the metallographic changes produced by flame cutting on various types of steel has shown that the characteristics of the steel materially affect the resulting microstructure. Not only carbon content and chemical composition but also the degree to which the austenitic grain size is controlled and the ability to deep harden are reflected in the structure of the flame cut edge. If the cutting flame is regulated to exaggerate these changes in place of minimizing them, as is ordinarily done, the structures produced in the steel may be used to determine some of the properties of the steel. This is the basis of flame testing.

A number of tests have shown that steel specimens may be graded by the metallographic appearance of the flame cut edge. When a properly shaped specimen is flame cut on one edge, five

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distinct zones are produced, each having a different metallographic structure. These five zones are:

1. *The normal zone* showing the original, precutting structure of the metal.
2. *A transition zone* produced by heating the metal just to the temperature of recrystallization which shows the manner in which gamma iron phase is formed from alpha ferrite and pearlite.
3. *A zone of small grain size* resulting from the transformation of a small-grained austenite.
4. *A martensitic zone* of large grain size resulting from the transformation of a coarse-grained austenite.
5. *A carburized zone* showing a hypereutectoid structure with needles of free cementite.

In addition to these five zones appearing in plain carbon steels, an austenitic zone is produced in certain alloy steels by the accumulation of the alloying element upon the cut edge. In such cases the austenitic zone replaces or obscures the carburized zone of carbon steels.

Standardized Conditions for Flame Testing—It is necessary to hold constant all variables in the cutting process if the effect of flame cutting is to be used to separate steels of different response to heat treatment. Two steels cannot be compared upon the nature of the cut edge if the structure of the edge is altered by variations in cutting speed, temperature or any of several other variables. These must be held constant. There are additional advantages in establishing uniform standards for any method of testing and these apply equally to flame testing. The standards set up for the present experimental work have proved satisfactory and it is suggested that they be adopted with certain indicated modifications.

In commercial flame cutting the common objective is to produce a smooth cut to exact dimensions with a minimum of effect on the metal surface and the greatest economy of gas, metal and time.

In flame testing the objective is to produce the greatest possible variation of steel structure under exactly controlled cutting conditions. In order to accomplish this, it is necessary to:

1. Use a standard type of cutting apparatus. (Hand controlled flame cuts cannot be used because of the impossibility of accurate control).
2. Produce the cut upon a standard shape and size of steel specimen.
3. Hold the temperature of the specimen before cutting to a room temperature range, 60 to 80 degrees Fahr.

4. Use a constant speed of cutting.
5. Use a definite ratio of preheat acetylene to cutting oxygen.
6. Use a standard cutting tip.
7. Examine the heat affected zone at a definite magnification.

The machines used for making oxygen cuts are becoming standard equipment in metal working shops. Nearly all of these machines are sufficiently flexible to accomplish the desired cutting conditions, smoothness of operation and ease of exact regulation.

Size and Shape of the Standard Specimen—The problem of selecting the most desirable shape and size of specimen for flame testing involves several variables. The depth of the cut produced in the steel determines, largely, the speed of cutting. Thus it is necessary to have sufficient thickness of metal to allow a relatively slow cutting speed. By slowing the cutting speed the time of contact between metal and flame will be greater and the corresponding effect upon the metal will increase. The thickness of the specimens tested in all of the experimental work was 1.25 inches. It is known, however, that a thickness of 1 inch will give the desired results and because of the saving of time in preparing the specimen this thickness should be adopted as standard.

The use of a sample having a width of at least 1.5 inches is of particular importance because of the quenching effect of the cold metal. Most of the metallographic structures revealed by flame testing are quench structures and if the results of different tests are to be compared, this quench effect must be constant. Rapid cooling of the heated edge occurs soon after the metal has been severed by the cutting flame. This rapid cooling originates from the flow of heat into the cool metal forming the bulk of the sample. If sufficient metal is not present to provide a reservoir for the heat from the cut edge, cooling will be slow and the metallographic structure of the edge will be altered. Quenching tests on plane carbon steel indicate that normal transformation of austenite to pearlite will occur when the heat of metal must flow a greater distance than 1 inch no matter what thermal gradient is produced at the quenched surface. This is then the limit of the useful width of the sample from the standpoint of quench effect. An additional 0.5 inch must be allowed for producing the cut, which makes the total width of the standard sample 1.5 inches. The length of the specimen is immaterial as long as the

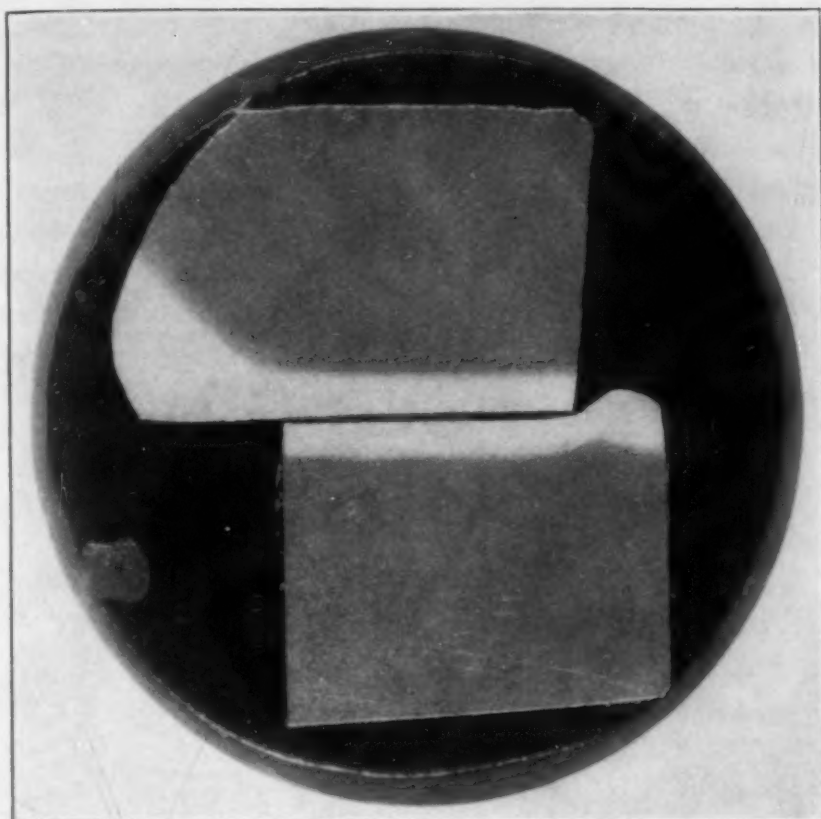


Fig. 1—Method of Mounting Samples for Metallographic Examination. $\times 3$.

metallographic sample is taken at least 1 inch from the start of the cut so that it will not be affected by the extra heating required to begin cutting. A total length of 3 inches is sufficient for the sample before flame cutting. A diagram showing the resulting steel sample and the proper place for removing the metallographic specimen is given in Figs. 1 and 2.

Standard Cutting Conditions—The effect of cutting speed has already been mentioned. All of the specimens used in this experiment were cut at a speed of 12 inches per minute, but for a sample 1 inch thick this is slower than is necessary. For the standard sample described above a cutting speed of 14 inches per minute is recommended. The diameter of the cutting orifice should closely approximate 0.005 inch. The smoothness of the bore of the cutting tip greatly affects the smoothness of the cut and to some extent the character of the edge produced on the steel. Oxygen pressure should be 30 pounds per square inch. Unfortunately, acetylene pressure varies according to the type of cutting equipment, and must be taken

from the manufacturer's recommendations. The cut edge should be smooth and free from irregularities caused by improper manipulation of the cutting machine.

Examination of the Cut Edge—The magnification at which the cut edge of the steel sample is examined is especially important as the structures produced by flame testing are much smaller than similar structures produced by furnace heat treatment. This necessitates using a magnification of one thousand diameters to determine grain size; a test that is ordinarily made at one hundred diameters. Simi-

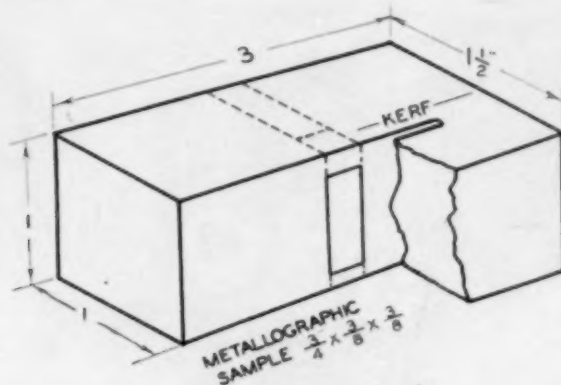


Fig. 2—Steel Sample for Flame Cutting Showing the Location of the Metallographic Sample.

larly, other structures in the steel are reduced in size to such an extent that several investigators of the metallurgical effects of flame cutting have stated that there is no zone of coarse grain size (12).¹ Actually there is a definite zone of comparative coarse grain size but it would be revealed as a fine grain zone when examined at the customary magnification of one hundred diameters.

Examination of the carburized outermost layer of the flame cut edge at one thousand diameters is difficult because of the tendency of the edge to become rounded during polishing. If the steel specimen is mounted in bakelite with the flame cut edge adjacent to another steel specimen as shown in Fig. 1, rounding of the edge is reduced to the point where a comparatively flat field may be obtained. Two different samples of steel may, in this manner, be polished and examined together; or one grade of steel may be flame cut normal and perpendicular to the direction of rolling and examined to determine the directional properties of the metal.

Results Obtained in Flame Testing—Any newly proposed meth-

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

od of metallurgical testing will require a large amount of study and correlation before it can be universally adopted. The data which have already been collected outline the general character of the test and indicate that the method has possibility of practical application. The photomicrographs in Figs. 3, 4 and 5 show the type of structure in each of the five zones produced in flame cutting together with the appearance of a sample of the steel from a McQuaid-Ehn test. From the appearance of the structure in these zones, it is possible to determine with more or less precision the following characteristics of a given steel.

1. Structure of the normal metal.
2. The ability of the steel to diffuse carbon uniformly when heated above the critical range.
3. The size of the austenitic grain at the time of its formation.
4. The aptitude of the austenitic grain for growth.
5. The ability of the steel to absorb and migrate carbon as expressed by McQuaid-Ehn normality or abnormality.
6. The presence of appreciable amounts of certain alloys such as nickel and manganese.

Microstructures Formed by Flame Cutting—The intense heating to which the cut edge is subjected during flame testing causes the steel to become austenitic. The transition between the normal steel, and the metal of the heated edge is clearly shown by metallographic examination. The second cut in each of Figs. 3, 4 and 5 shows the nature of this transition zone. The tendency of the grains of both ferrite and pearlite to break up into smaller grains is marked. The pearlite has lost its characteristic laminated structure indicating that it was partially transformed into austenite and that the metal in this transition zone was heated just above the Ac_1 critical point. In the case of the slow heating in ordinary heat treating operations it is usual to find little change in the size of the ferrite grains until the metal has been heated close to or above Ac_3 and held until most if not all of the metal has formed austenite. Upon cooling, the excess ferrite is forced out of solution on the boundaries of the austenitic grains with the result that there is often a refinement of the grain size of the steel. In flame cutting the metal in the transition zone has apparently gone through this same cycle but such has been the rate of heating and cooling that the refinement of grain size is greatly intensified.

The third cut for each steel shows a martensitic structure result-

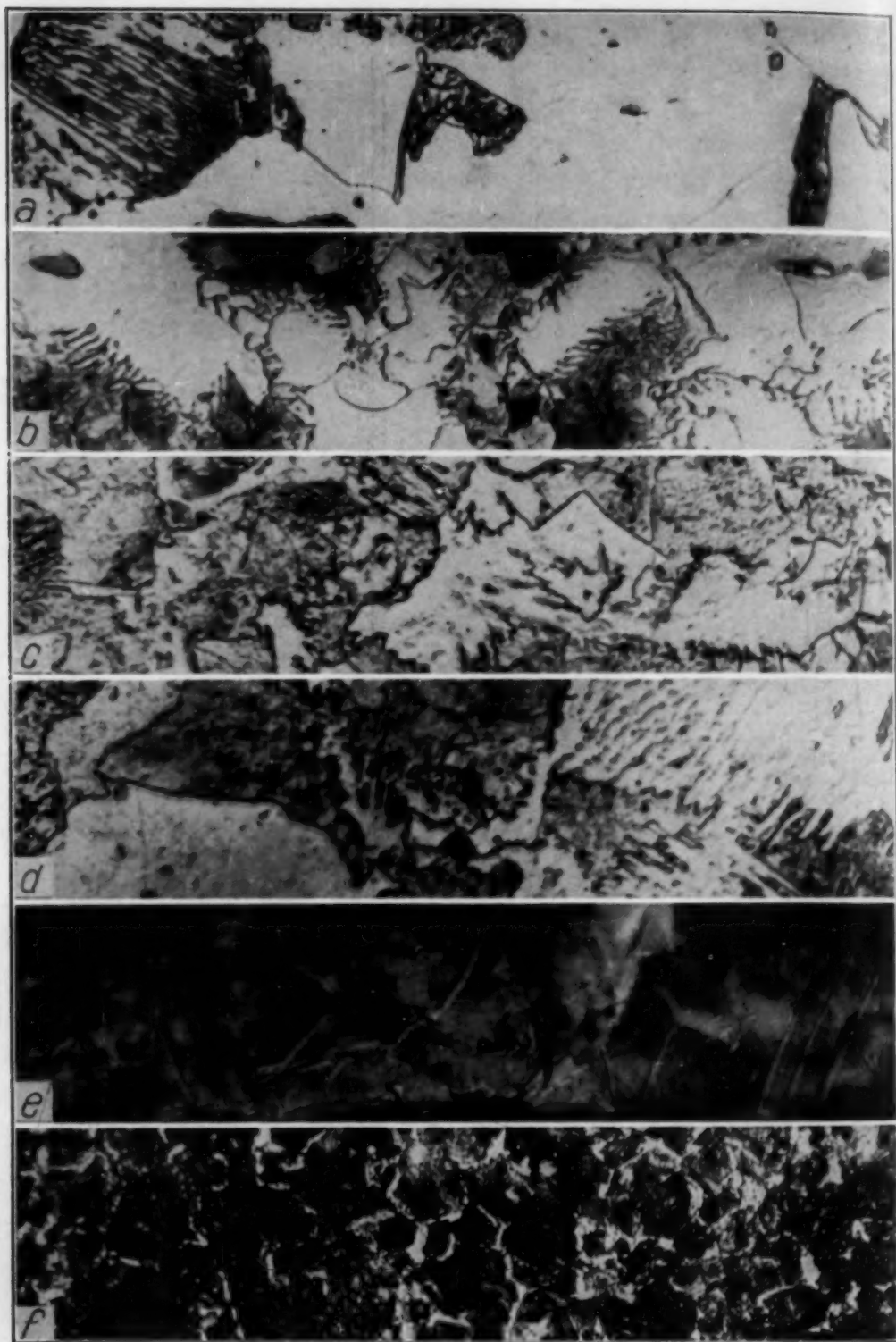


Fig. 3—Photomicrograph of Structures Revealed in Steel No. 20. (S.A.E. 1020).
a. Normal Metal. $\times 1000$. b. Flame Cut Edge—Transition Zone. $\times 1000$. c. Flame Cut Edge—Small Austenitic Grain. $\times 1000$. d. Flame Cut Edge—Large Austenitic Grain. $\times 1000$. e. Flame Cut Edge—Carburized Zone. $\times 1000$. f. McQuaid-Ehn Carburized Zone. $\times 100$. (Carburized at 1700 Degrees Fahr. for 8 Hours).

ing from the completion of transformation and the subsequent rapid cooling of the metal. The grain size represented is the same as the austenitic grain size at the time of its formation. The magnitude of this crystal size is important as it represents the size of grain that may be attained by extremely rapid heating followed by rapid cooling. Grain size greatly influences the physical properties of steel, and it may be possible to predict some of the property changes resulting from annealing or other heat treatment processes by correlation with the size of the nascent austenitic grain as shown by flame cutting.

The growth of the austenitic grain is shown by the size of the martensitic grain appearing in the third cut of Figs. 3, 4 and 5. The rate and temperature at which the austenitic grain growth takes place is not the same for all steels. In certain types of steel the austenite is sluggish and does not tend to increase in grain size as temperature above the upper critical point is raised. Such steels are known as "controlled grain steels" and are usually produced by adding aluminum to the molten metal while it is in the ladle. The aluminum reacts with the oxygen in the metal and forms aluminum oxide, but whether the effect of aluminum in producing fine grain is due to the dispersed particles of oxide or to the aluminum in solution in the steel is still a controversial point.

The chief point of interest from a testing standpoint is the differentiation of "controlled grain steels" from the steels whose austenitic grains grow normally. This determination resolves itself into a problem of degree as all steels will form small or large grains depending upon the temperature to which they are heated and the time at the elevated temperature. In the case of "controlled grain steels", the flame test shows a wider zone of small grain size at the expense of the zone of coarse grain size, and as the width of this zone is not marked by definite boundaries, it is difficult to determine quantitatively the extent to which the grain is controlled. Quantitative determination is usually not important because "controlled grain steels" are nearly always made in such a manner that the grain will be as small as possible while the coarse-grained steels, on the other hand, show little tendency to retain the fine austenitic grain when heated much above the critical range.

Diffusion of Carbon and Absorption of Ferrite—Ordinarily commercial steel, containing less than 0.25 to 0.35 per cent carbon, will form martensite only with the most drastic quench and will exhibit

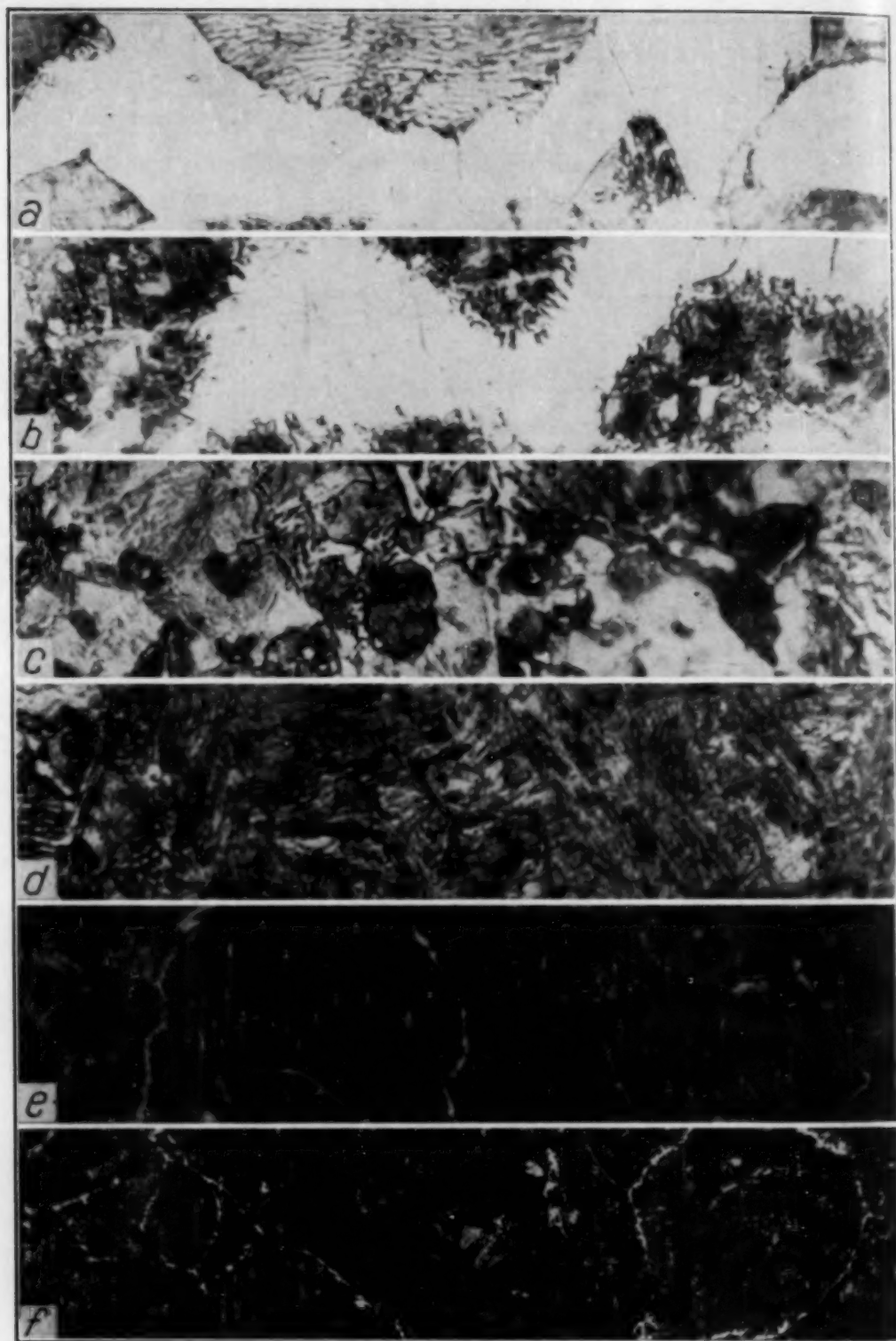


Fig. 4—Photomicrograph of Structures Revealed in Steel No. 25. (S.A.E. 1035).
a. Normal Metal. $\times 1000$. b. Flame Cut Edge—Transition Zone. $\times 1000$. c. Flame Cut Edge—Small Austenitic Grain. $\times 1000$. d. Flame Cut Edge—Large Austenitic Grain. $\times 1000$. e. Flame Cut Edge—Carburized Zone. $\times 1000$. f. McQuaid-Ehn Carburized Zone. $\times 100$. (Carburized at 1700 Degrees Fahr. for 8 Hours).

a great tendency to expel ferrite from solution or will not take all of the ferrite into solution when the steel is heated above the upper critical point. This is especially true of steels containing large amounts of oxide and those that are killed with aluminum to form controlled steel. Since low carbon steels are generally high in oxygen content, it is commonly assumed that these steels are not suitable for heat treatment. The division between steels that deep harden satisfactorily upon quenching from a temperature above the upper critical point and the steels that do not harden in this manner cannot be made upon a basis of carbon content alone as it is found that some low carbon steels are susceptible to heat treatment and harden greatly when quenched, while other medium carbon steels do not. The ability to deep harden and to increase the physical properties that accompanies hardening depends largely upon the ability of the steel to dissolve all of the ferrite in the austenitic solid solution and to hold the ferrite uniformly dispersed in the form of martensite when the steel is quenched. A steel possessing this characteristic is called a "regenerating" steel.

Ability to absorb and retain all of the free ferrite in the austenite is measured in flame testing by the appearance of the grain boundaries of the martensitic zone particularly of that part of the heat affected edge which attains a large austenitic grain. If the steel is completely regenerating, it will form uniform martensite in this part of the heat affected zone as is shown in the fourth specimen of Fig. 4 for steel No. 25. If the steel is nonregenerating, it will have free ferrite along the edge of the martensitic grains. In flame testing the temperature of the edge of the steel may approach the melting point and under these conditions it is possible for almost any type of steel to absorb all of its free ferrite in the austenitic solution. When this type of steel is cooled from these high temperatures, it tends to throw the ferrite out of solution before normal transformation at the critical range can occur. As the austenitic solution from which the ferrite is discharged is in the gamma crystalline form, the free ferrite thus discharged from solution has a characteristic appearance greatly resembling some of the structures that are grouped under the term "Widmanstätten" structure. The appearance of this saw tooth structure is shown by the fourth specimen in Fig. 5 for steel No. 45.

The McQuaid-Ehn Test and Flame Testing—The ability of carbon to migrate in steel when heated above the critical range is made use of in the McQuaid-Ehn test to determine some of the

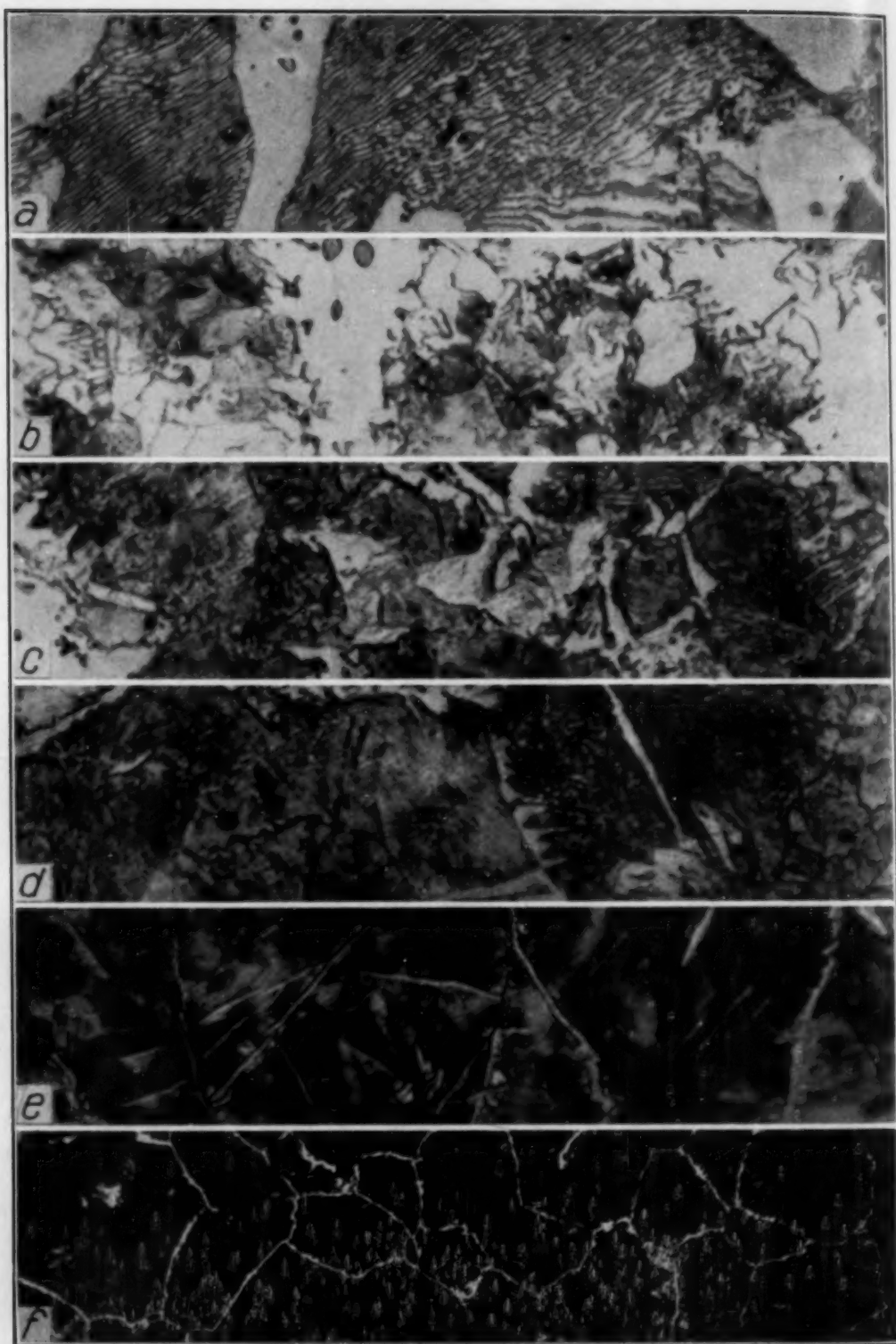


Fig. 5—Photomicrograph of Structures Revealed in Steel No. 45. (S.A.E. 1040).
a. Normal Metal. $\times 1000$. b. Flame Cut Edge—Transition Zone. $\times 1000$. c. Flame Cut Edge—Small Austenitic Grain. $\times 1000$. d. Flame Cut Edge—Large Austenitic Grain. $\times 1000$. e. Flame Cut Edge—Carburized Zone. $\times 1000$. f. McQuaid-Ehn Carburized Zone. $\times 100$. (Carburized at 1700 Degrees Fahr. for 8 Hours).

properties of the steel in the austenitic state. Originally the test was employed to select steel for case carburizing purposes, but its use has been extended to the determination of the austenitic grain size and to the reaction of the steel to heat treatment in general. The steel to be tested is heated in a special carburizing compound to 1700 degrees Fahr. (925 degrees Cent.) and held at that temperature for sufficient time to produce a hypereutectoid case on the surface of the specimen. Ordinarily the time at this temperature is 8 hours, although this may be shortened under certain conditions. The specimen must be cooled slowly to permit the formation of pearlite and allow the excess cementite to accumulate at the grain boundaries.

The appearance of the splines of cementite on the grain boundaries is found to be narrow and continuous in normal steels and the grains will be completely pearlitic. Abnormal steels exhibit thickened splines and accumulation of cementite in masses with discontinuous formation of the cementite splines. The pearlite tends to be coarse and irregularly laminar and breaks down at the grain boundaries into masses of cementite and ferrite—a condition that causes the formation of soft spots on case carburized articles. The gradation between normal and abnormal steels is continuous so that any degree of normality may be observed.

There appears to be a definite relationship between normality and the grain size of the hypereutectoid zone of the carburized flame cut edge. Steels that form coarse grains in this zone are found to be “normal” while the fine-grained steels are nearly always “abnormal”. Cases of fine-grained normal steel are occasionally reported. It is probable that the determining factors as to both normality and grain size originate in the melting practice and may be independent. Steels with controlled austenitic grain are ordinarily made in such a way that they give an abnormal reaction to the McQuaid-Ehn test.

The absorption and migration of carbon is revealed in flame testing by the appearance of the carburized zone. A hypereutectoid zone is formed on the edge of the steel by the action of the cutting flame, and during cooling the excess cementite is forced to the grain boundaries or deposited along the cleavage planes of the austenitic grains in a characteristic triangular pattern. Cooling is too rapid to permit the formation of normal pearlite, but the appearance of the cementite splines in flame cut carburized zone is strikingly similar to their appearance in the McQuaid-Ehn test. This is especially true of the normal steels as shown by Figs. 4 and 5. In each figure the

bottom cut shows the appearance of the hypereutectoid zone formed in the McQuaid-Ehn test while the cut just above shows the appearance of the carburized zone formed during flame testing. The degree of magnification for comparison is not the same in the two tests

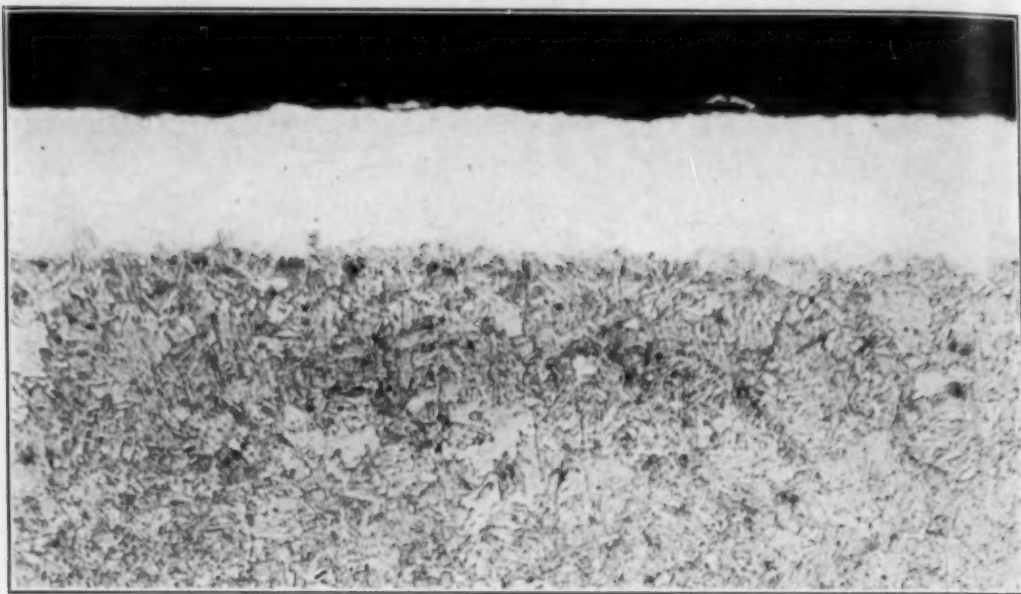


Fig. 6—Edge of Flame Cut Nickel Steel No. 320. The Light Colored Zone Along the Edge is Austenite which is Prevented from Decomposing by its High Nickel Content. $\times 125$.

as the flame cut edge has a much finer-grained structure and must be examined at one thousand diameters as compared with one hundred diameters commonly used in the McQuaid-Ehn test.

Flame Testing Alloy Steels—The character and physical properties of steel are greatly altered by the addition of a few per cent of

Table I
Chemical Analysis of Steels Tested

Sample No.	Carbon	Manganese	Phosphorus	Sulphur	Nickel
20	0.20	0.36	0.016	0.035
25	0.34	0.58	0.006	0.034
45	0.42	0.57	0.008	0.040
320	0.20	0.40	0.014	0.031	3.47

such alloying metals as nickel and chromium. The micro-structures are distorted, particularly the structures resulting from rapid cooling through the critical range. The effect of these alloys upon the structures resulting from flame testing is similar to the effect produced by conventional heat treating. Evidence as to the character of the

steel must be weighted with an allowance for the alloy factor. Alloy steels normally show a greater depth of heat affected zone when flame cut. The martensite produced is finer and ordinarily more uniform. The austenitic grain size cannot be determined with the same accuracy that is achieved in plain carbon steel.

The most striking effect of alloy additions on the flame cut structure is the development of an austenitic zone on the edge of the test piece. When alloys such as nickel are present, there is an actual increase in alloy content of the steel along the flame cut edge. This increase in alloy content combined with an increase in carbon content and the rate of cooling to which the edge is subjected allow the retention of austenite at room temperature. The structure of this zone, for a nickel steel, is shown in Fig. 6. The light colored coating of austenite is approximately 0.007 inch in thickness. It is backed up with several hundredths of an inch of martensite. A number of tests were made to check the formation of austenite in this manner. Etching with dilute copper sulphate showed no attack upon the austenitic layer although all other surfaces of the metal were affected. Analysis of metal from the flame cut edge showed an increase of over 300 per cent in nickel content while an analysis of the slag produced after correction for the weight of the oxygen showed a loss of nickel.

Manganese and silicon steels react in a manner similar to the nickel steels as do some other alloys, especially those of the solid solution forming types. When a white austenitic zone is formed in flame testing, it is due to the presence of appreciable amounts of alloying metals. This provides a convenient method for checking the presence of some alloys with the metallograph even when these alloys are present in amounts as low as 1 per cent or less in the steel.

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THE INFLUENCE OF MAGNETIC FIELDS ON DAMPING CAPACITY

By E. R. PARKER

Abstract

Damping tests were made on two commercial steels in steady and alternating magnetic fields. The energy loss when a steady field was present was found to be less than half the value for zero field. An alternating field reduced the energy loss even further, to about half the value obtained in a steady field. The endurance of fatigue specimens was found to be reduced by the presence of a steady magnetic field.

DAMPING capacity is defined as the energy dissipated as heat by a unit volume of metal during a completely reversed cycle of stress. The specific damping capacity, P , generally used in damping measurements, is the per cent of the total energy absorbed per cycle.

The mechanism of this internal absorption of energy is only partially understood. Damping capacity is undoubtedly composed of several components, each of which contributes its part to the total energy absorbed. One source of energy loss in a freely vibrating bar results from the heat flow along a stress gradient produced by a rapidly applied load. A sudden adiabatic stretching of a bar causes its temperature to drop, and conversely sudden compression causes a rise in temperature. When a bar is bent, the side in compression is slightly heated and the side in tension is slightly cooled. The heat will tend to flow from the higher to the lower temperature and thus produce a loss of elastic energy. A vibrating bar will be subject to such an energy loss and the magnitude of this loss will depend upon the material, the vibration frequency, and the dimensions of the test bar. Zener (1)¹ has made some calculations which show that this type of energy loss can be large under certain conditions.

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

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A slight plastic yielding, even within the so-called elastic limit, furnishes its contribution to the energy loss. This has been demonstrated by extremely accurate stress-strain measurements, which show a hysteresis loop even at very low stresses (2).

Another component of energy loss, which is perhaps the most interesting, is the magnetic or electrical loss. It has been known for some time that a vibrating bar of ferromagnetic material will have a smaller damping capacity when tested in a magnetic field (3), (4). The mechanical and magnetic forces are so related that the presence of a magnetic field will cause a bar of ferromagnetic material to change its length—this phenomenon is known as magnetostriction. Conversely, if such a bar is stretched by a mechanical force, the magnetization of the bar is changed. If the direction of the stress is reversed, the magnetization again changes, so that an alternating stress will cause a continuous change in magnetization and local eddy currents will be produced within the bar which will add to the energy loss. A strong magnetic field prevents the mechanical force from changing the magnetization and the eddy current loss is eliminated. This explanation seems to fit the data for steady magnetic fields very well (4), (5), (6).

Some experiments were made along this line by the author to find the magnitude of this effect for some commercial steels. Since no work seems to have been reported in the literature on the effect of alternating magnetic fields as contrasted with steady fields, the experiments were extended to cover this case. The results, reported herein, are very interesting because they show that an alternating field is even more effective than a steady one in reducing the energy loss.

DESCRIPTION OF APPARATUS

The damping was measured in a Foeppel-Pertz machine (2), (7) which had been altered slightly for the magnetic tests. The ends of the cross-arm and the pole pieces which were added to the frame were made arcs of two concentric circles of proper diameter so that the air gap between them would be constant for all positions of the vibrating cross-arm. The electromagnets used to hold the cross-arm in the strained position before a test were replaced by a nonmagnetic holder. These changes eliminated the small variation in reluctance of the magnetic path which occurred during the oscillation of the

cross-arm in the original machine. The magnetic field was produced by a solenoid concentric with the specimen. The lines of flux were parallel to the axis of the specimen, which was subjected to reversed torsional stress. The machine is shown in Fig. 1.

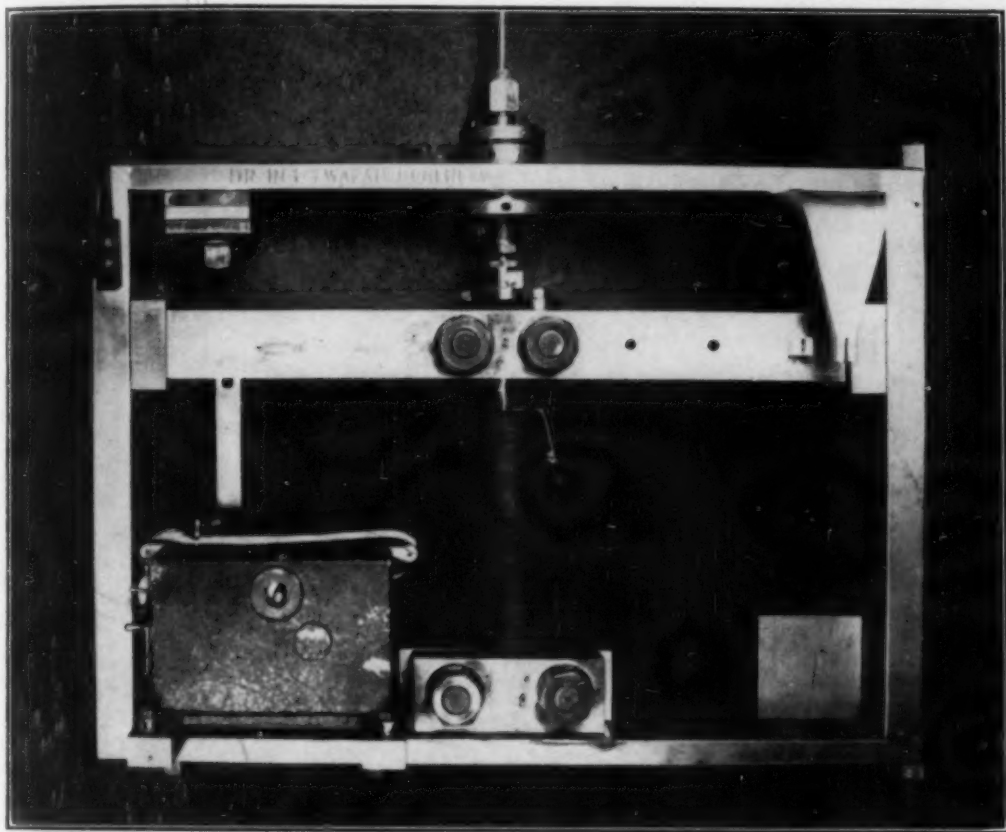


Fig. 1—Foeppel-Pertz Damping Machine Showing Pole Pieces, Nonmagnetic Release and Solenoid.

The magnetic calibration of all specimens was carried out in the damping machine so that the current through the solenoid could be converted directly to flux density.

With the method of measuring damping used in these experiments, the results could be duplicated to within one part in twenty-five.

TEST RESULTS

One of the specimens tested was a 0.40 per cent carbon steel which had been air cooled from 850 degrees Cent. (1560 degrees Fahr.) and drawn at 650 degrees Cent. (1200 degrees Fahr.). This bar was tested in various steady and alternating fields. In Fig. 2

the damping-stress curves have been plotted for no field, a high steady field and a high alternating field. The energy loss is much smaller for the alternating field than for the steady or zero fields. The aver-

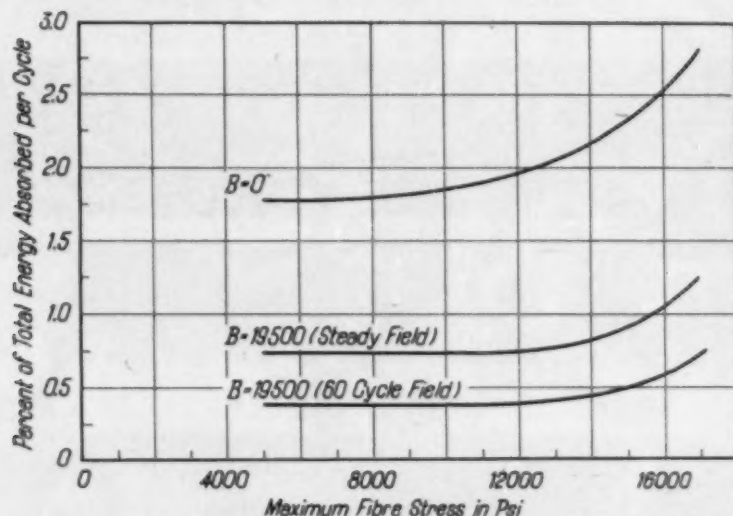


Fig. 2—Damping-Stress Curves for a Normalized 0.40 Per Cent. C Steel at Zero Field, a High Steady Field, and a High Alternating Field.

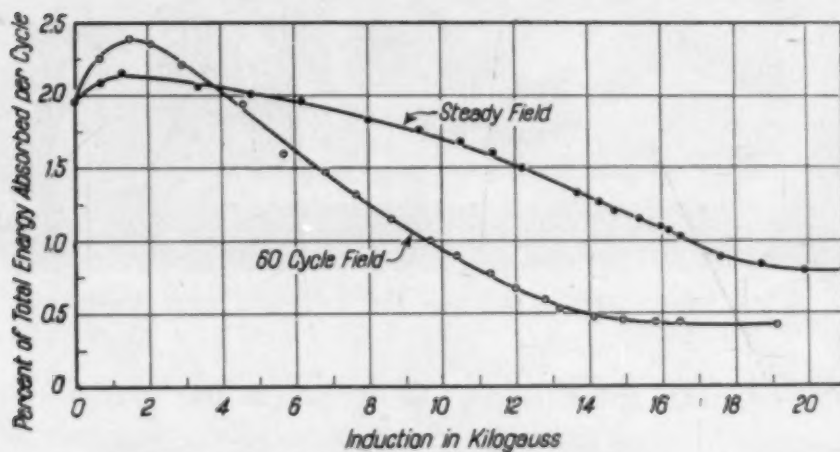


Fig. 3—Damping-Induction Curve for Normalized 0.40 per Cent Carbon Steel. The Average Damping for Stresses Between 8500 Pounds Per Square Inch and 17,000 Pounds Per Square Inch Has Been Used for the Ordinate.

age damping capacity between stresses of 8500 pounds per square inch and 17,000 pounds per square inch (maximum fiber stress) has been used for comparing the damping in various fields. The results obtained with direct current and 60-cycle current are shown in Fig. 3. It is interesting to note that as the field increases from zero, the damping at first increases, then decreases until the steel becomes saturated, and then remains constant.

The specimens had to be completely demagnetized before each test, because the value of the flux density corresponding to a given magnetizing current was known only for points on the magnetization curve and not for the hysteresis loop. The effects of 40, 25, 15 and 6-cycle fields are now being investigated (the natural frequency of the machine is 10.5 cycles per second). Preliminary tests indicate that as the frequency of the field decreases the damping gradually increases.

A specimen of the 0.40 per cent carbon steel was quenched to give maximum hardness and then tested in steady and alternating fields. The damping was very low in all cases and the magnetic fields had practically no effect.

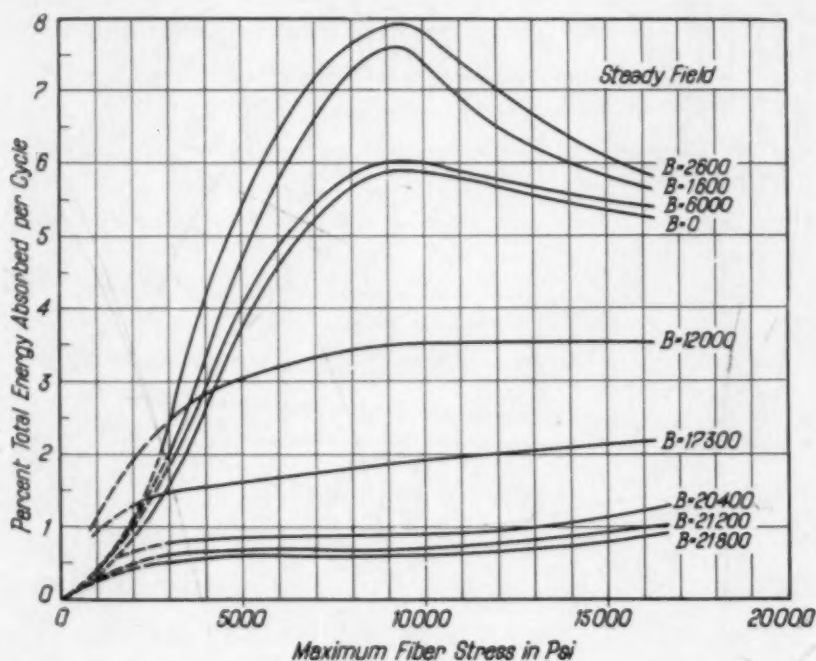


Fig. 4—Damping-Stress Curve for Annealed 5.0 Per Cent Chromium, 0.5 Per Cent Molybdenum, 0.20 Per Cent Carbon Steel in Various Steady Fields.

The damping curves shown in Fig. 2 are typical of those obtained for most materials. There are certain alloys, particularly the high chromium-irons, which behave somewhat differently. The damping-stress curves, such as those shown in Fig. 4 for an annealed 5.0 per cent chromium, 0.5 per cent molybdenum, 0.20 per cent carbon steel, exhibit a peculiar maximum which increases with low fields but entirely disappears at high fields. As shown in Fig. 5, the same phenomenon occurs to an even greater degree when a 60-cycle field

is used instead of a steady one. These experiments show that the peak in the damping-stress curve is due to the magnetic component of damping.

It is important to note that the effect of magnetic fields on damping is different at different stresses. At high stresses, the damp-

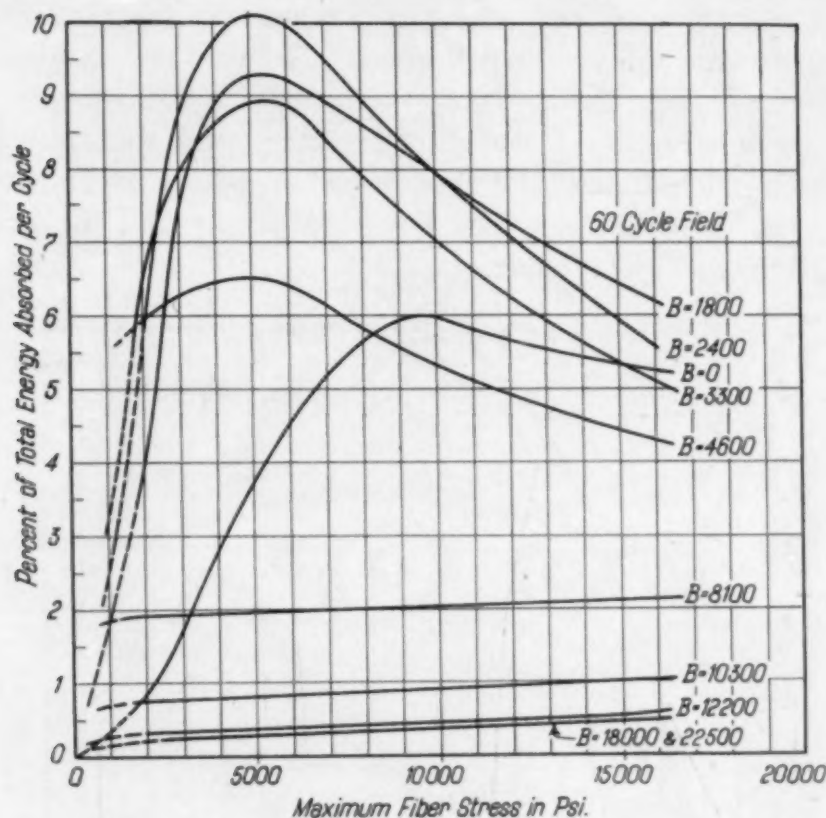


Fig. 5—Same as Fig. 4 in Various Alternating Fields.

ing is greater at low fields than at zero field, but at higher fields the damping is smaller. This is reversed at low stresses, where the damping is greater in strong fields than in weak fields. It becomes apparent from this that damping studies should include the variable stress whenever fundamental data are required.

Some important work by Brophy (8) has shown the possibility of a correlation between the damping capacity and the sensitivity to stress concentration of a material subjected to fatigue stresses. In view of this work it seems possible that a magnetic field may affect the notch sensitivity of a material. Some tests were made on the 0.40 per cent carbon steel used for the damping experiments. This material had an endurance limit of 39,000 pounds per square inch in the absence of stress concentrations and 21,000 pounds per square

Table I
Cycles to Failure

Stress Lbs. Per Square Inch	No Field	Steady Field (300 Oersteds)	60-Cycle Field (2.5 Oersteds)
22,500	1,432,500	715,700	1,712,300
22,500	1,589,800	843,600	1,724,700
20,500	13,053,500*	3,339,100	4,496,800
20,500	34,000,000*	4,600,000
20,000	4,431,300
20,000	7,213,600
19,500	12,000,000*	4,933,100	18,167,000*
19,500	5,363,000
18,500	8,067,100
17,500	15,000,000*

*Indicates unbroken specimen.

inch with a 3:1 theoretical stress concentration. Both types of bars were tested in a steady field of 300 oersteds and a 60-cycle field of 2.5 oersteds. The high steady field could be used with no appreciable overheating of the bar, but the alternating field had to be kept very low to prevent appreciable heating of the bar by eddy currents. The data obtained on notched test bars, which were made as nearly alike as possible, are listed in Table I. Although these data are not too conclusive, it is apparent that the presence of a steady magnetic field adversely affects the endurance of a notched bar. The endurance of the bars having no stress concentration was not affected by the magnetic field.

SUMMARY

Damping tests were made on a 0.40 per cent carbon steel and a 5.0 per cent chromium, 0.5 per cent molybdenum steel in steady and alternating magnetic fields. The energy loss when a steady field was present was less than half the value for zero field. An alternating field reduced the energy loss still further, to about half the value obtained in a steady field.

The maximum in the damping-stress curve for high chromium-iron alloys was found to be due to the magnetic component of damping. The importance of measuring damping at different stresses is demonstrated by the results for the 5.0 per cent chromium, 0.5 per cent molybdenum steel where the effect of magnetic fields on damping is different at different stresses.

The endurance of fatigue specimens having stress concentrations was found to be reduced by the presence of a steady magnetic field.

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DISCUSSION

A. V. DEFOREST:¹ This is a most entrancing subject. There is something very queer about magnetism, anyway, especially alternating, and there is something more than that about damping, and when the two get mixed up, the results are plainly evident.

There is, however, a very old and very interesting relationship between these two which has been observed the other way around. For a long time people have measured magnetic hysteresis loops and many years ago the effect of slight mechanical vibrations on the magnetic hysteresis loop was observed.

Mr. Parker is studying the effect of the magnetic field superimposed upon a mechanical hysteresis loop, and the result is somewhat of the same variety, perhaps, if not exactly in detail the same thing, as occurs when a mechanical stress cycle is applied to a magnetic hysteresis loop. In both cases the loop is reduced in area.

It is quite evident that because a magnetic field sets up a mechanical stress and also a mechanical change in dimensions, that what we are doing with the alternating magnetic cycle is to superimpose on a slow mechanical cycle a high speed magnetic cycle which may be operating through its stress due to magnetostriction rather than through any mysterious motion of the magnetism. There may be a secondary indirect magnetic effect but there is certainly also a direct mechanical effect due to the magnetic cycle.

To get at this thing, we would have to know much more than is now particularly evident about the magnetostrictive relationship.

It is interesting that the tests shown here, carried out by Mr. Parker, show a longitudinal magnetic field and a torsional mechanical stress superimposed. The longitudinal magnetic field produces a longitudinal change in

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dimensions, primarily, while the torsional stress is at right angles as far as the shear is concerned.

Circular magnetization, producing circular magnetostrictive effects, might also be expected to produce large disturbances in torsional vibration.

I might add that a similar result occurs in the effect of a magnetic cycle on primary creep. If a piece of steel is loaded in torsion and is creeping logarithmically and an alternating current field is superimposed, that creep is immediately reduced and wiped out almost to zero; so that there again is an indication of the powerful relation between these inelastic mechanical effects and the magnetic field.

J. T. NORTON:² I certainly agree with everything that Professor deForest has said about the complexity of damping and the difficulties of interpretation. However, Mr. Parker has brought out a point which I would again like to emphasize, that is, the importance of qualifying damping capacity, not by a numerical value but by a curve, because very different things happen at low stresses than happen at high stresses.

The other thing in his paper which I would like to comment upon is that he mentioned that these particular alloy steels show a peak in the damping curve. The damping rises rapidly to a maximum and then falls off again. That same behavior is characteristic of Armco iron and the low carbon steels which have been annealed, and they show this anomalous behavior very plainly indeed. That peak in the damping curve is wiped out in much the same way that it is wiped out by a magnetic field if a small amount of cold work is put in the sample. If it is twisted in torsion so it is deformed plastically, that peak in the damping curve seems to disappear, but it comes back again as soon as the material has been annealed. It is a perfectly reproducible cycle.

R. S. DEAN:³ We have done considerable work lately on manganese alloys which have extremely high damping capacities at low stresses, wherein damping capacities of the order of 10 or 11 per cent were determined. From a theoretical standpoint, I am inclined to think that working at low stresses certainly simplifies matters a lot.

Just as a matter of interest, we made a few measurements bearing on Professor Norton's comments. We have not made enough of them to discuss them quantitatively, but it happens that some time ago we studied the relation between torsional work and coercive force in a number of magnetic materials. The results were published in one of our rather obscure mimeographed Bureau of Mines publications, so probably none of you saw it. It reported some very careful work on the relation between torsional work and coercive force in four materials, viz., pure electrolytic iron furnished by Westinghouse Electric and Manufacturing Company; some Permalloy from the Bell Telephone Laboratories; some very pure nickel from the International Nickel Company; and some nonaging Armco.

Those four materials were all very carefully annealed; in the case of the Armco, I believe Professor Chipman, then with Armco, annealed it for us in some mysterious way that I do not know, but in a way that was supposed to

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³Metallurgist, U. S. Bureau of Mines, Salt Lake City, Utah.

render it entirely nonaging, so that we have measured the change of coercive force with work by taking a whole series of rods and giving them varying torsional work. Those relationships I would not go into, but when we got interested in damping capacity, we went back and took those rods, which were worked in measurable amounts by very low stresses, and we found that the nonaging Armco was entirely different from the others. The others dropped off very sharply in damping capacity with a small amount of cold work. It went down and stayed down. That is true of the pure nickel, of Permalloy and of pure iron. The nonaging Armco, on the other hand, went through a minimum and then went back up.

So it would appear that for a simple magnetic material the damping capacity falls off very rapidly with very low amounts of cold work; that is, the damping capacity measured at very low stresses.

Author's Reply

As the discussers have so ably pointed out, damping is indeed a complex property, and it will probably be some time before it is completely understood.

Professor deForest has discussed the possibility of a relationship between magnetostriction and damping. One experiment which I performed along this line may be of interest. A six per cent silicon iron is known to have essentially zero magnetostriction, i. e., no change in mechanical dimensions occurs when the alloy is magnetized. If the magnetic field has no effect upon the mechanical strain, a mechanical strain would normally be expected to have no effect upon the magnetic properties. Thus the damping of the six per cent silicon iron alloy might reasonably be expected to show a different behavior from other alloys when tested in various magnetic fields. However, the tests disclosed that this alloy acted exactly like the others. The damping was decreased to about half by the presence of a strong steady field and to about one-quarter by an alternating field. It may be that in this alloy there is an anisotropic magnetostrictive effect in each single grain of the polycrystalline bar, which is of such a nature that each grain would change dimension when magnetized, but the summation of the dimension changes over the length of the bar would be zero.

BRIGHT HARDENING OF TOOL STEELS WITHOUT DECARBURIZATION OR DISTORTION

By J. R. GIER AND HOWARD SCOTT

Abstract

This paper describes an improved method of bright hardening tool steels without surface damage by oxidation and decarburization. The method applies particularly to the alloy steels of the air hardening type used for blanking dies. These steels are fully hardened in large sections without distortion by natural cooling in a protective atmosphere. A specially designed furnace is essential for the application of this process. The development of suitable equipment and protective atmosphere is described here.

DAMAGE to tool steels during hardening by oxidation, decarburization, distortion or cracking or some combination of these effects is widely accepted as inevitable. Though avoidance of all four types of damage offers considerable difficulty, this objective is attainable when the steels under consideration are limited to the air hardening type. Within that important class of steels are a wide variety of compositions suitable for practically all kinds of tool applications. This paper describes a new means for bright hardening them by gas cooling within an atmosphere that eliminates oxidation and decarburization from the hardening equation.

Distortion in the sense used here is the irregular or unpredictable changes in dimensions or contour in excess of the normal increase in specific volume which always accompanies the hardening of steel with the production of martensite. Both distortion and cracking are avoided in this procedure by cooling in the protective atmosphere. The cooling rate is sufficiently fast nevertheless so that full hardening of air hardening steels is obtained in solid blocks at least 2 inches thick without grain coarsening.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, J. R. Gier is associated with the chemical and metallurgical division, and Howard Scott is section engineer, metallurgical section, Westinghouse Research Laboratories, E. Pittsburgh, Pa. Manuscript received July 7, 1939.

Oxidation is now frequently avoided by heating for hardening in one of the commercial protective gas atmospheres, but serious decarburization can still occur even though oxidation is prevented. The widely used air hardening tool steels must be heated for hardening to or above 1700 degrees Fahr. (925 degrees Cent.) at which temperatures decarburization is extremely rapid, due more to the high temperature required for carbide solution than to any direct effect of alloy content. To allow for this damage as well as for distortion, it is common practice to leave extra stock which must be removed after hardening by grinding. Finish grinding, as well as the hand stoning required to finish fine interior contours, adds greatly to the cost of dies and tools.

On investigating furnace atmospheres one was found which prevents decarburization as well as oxidation and is readily available. Further development, however, was required to convey this gas unimpaired at furnace temperatures to the surface of the work. The equipment must then meet several basic requirements. There must be provisions for heating and cooling the work within the furnace atmosphere and for moving the work without contamination of the atmosphere as by opening doors. A satisfactory atmosphere must be delivered to the heating chamber and remain innocuous to the hot work while in contact with it. The work must be so supported as to heat and cool uniformly and freely but without distortion from sagging under its own weight. Means for doing these things are described in following sections of this paper.

PROTECTIVE GASES

The purpose of protective atmospheres is to prevent those chemical reactions between hot steel and its environment that cause scaling and undesirable changes of carbon content. To prevent scaling it is necessary that the oxygen pressure in the atmosphere be less than the dissociation pressure of the oxides of the metal being heated; this is most readily accomplished by the presence of a reducing gas such as hydrogen or carbon monoxide. The proportion of total oxidizing impurities that may be present with maintenance of reducing conditions is governed by the stability of the oxides of the particular metal. The major oxidizing impurity in hydrogen is H_2O while those in combusted gas are H_2O and CO_2 . Steels of high chromium content are oxidized by very small proportions of these impurities while car-

bon steel is not oxidized by relatively high concentrations of them.

Reactions causing changes in surface carbon are controlled by an entirely different set of conditions from those affecting oxidation and reduction and must therefore be considered separately. Hot steel may be regarded as exerting an immeasurably small but definite carbon pressure that varies with temperature and with the amount of carbon in solid solution. Gas mixtures containing carbon monoxide or hydrocarbons may also be regarded as having a similarly minute carbon pressure or carbon-forming potential. The actual magnitude of this carbon pressure in gas mixtures at equilibrium is controlled by the relative amounts of the active components according to the law of mass action.

For example, in a mixture containing CO and CO₂ in equilibrium, the carbon pressure may be expressed

$$[C] = k_1 \frac{[CO]^2}{[CO_2]}$$

Similarly in a mixture of H₂ and CH₄ the carbon pressure may be expressed

$$[C] = k_2 \frac{[CH_4]}{[H_2]^2}$$

The bracketed terms denote partial pressures and the terms k_1 and k_2 are the equilibrium constants of the respective reactions, which also vary with temperature.

When the carbon pressure of hot steel is less than that of its surrounding gas, carbon will be lost from the steel to the atmosphere, and vice versa. The rate at which this transfer occurs, which is of utmost practical importance to the heat treater, depends in part upon the magnitude of this difference in carbon pressures between the steel and the gas. This rate increases rapidly with increase in temperature; it is different for different steels and varies with the amount and kind of active gas components present. For example a pure dry hydrogen atmosphere has zero carbon pressure and accordingly will theoretically decarburize steel by the reaction $(Fe, C) + 2H_2 \rightleftharpoons CH_4 + Fe$ but this reaction is extremely slow as compared with the reactions $(Fe, C) + CO_2 \rightleftharpoons 2CO + Fe$ and $(Fe, C) + H_2O \rightleftharpoons H_2 + CO + Fe$.

There are three basic types of protective atmosphere suitable for heat treating tool steels without surface damage. These are (1) mix-

tures containing 20 to 40 per cent CO which tend to be actively carburizing or decarburizing unless their CO_2 content is closely regulated, (2) chemically pure H_2 which is actively reducing to all steels and decarburizes only at a very slow rate, and (3) chemically pure N_2 which is totally inert so far as oxidation and decarburization are concerned.

In selecting one type atmosphere for tool-hardening from this group the primary considerations were:

- (1) The gas should be capable of protecting all steels against loss or gain in surface carbon and not require adjustment of composition to suit various steels.
- (2) The gas should be strongly reducing to permit heating high chromium steels without oxidation.
- (3) The gas should be simple to prepare and not require elaborate processing.

Active gases containing high CO, such as may be obtained from a charcoal gas producer, were rejected because their carbon pressure and reducing power are both dependent on the CO- CO_2 ratio and therefore cannot be separately controlled. This characteristic is not objectionable in treating plain carbon and low alloy steels for which the reducing power of this type of gas is adequate over the entire working range of carbon pressures. However, for high chromium steels of low carbon content this gas is unsuited for bright hardening as it must be made highly reducing to prevent tarnish and this would require a high CO- CO_2 ratio such that the gas would be actively carburizing.

Nitrogen lacks the reducing character needed in a protective atmosphere; moreover, it is expensive when purchased in bottles and must be further purified. When it is prepared from combusted gases it is much cheaper, but expensive processing equipment is still required. This leaves pure hydrogen as the most favorable type of gas atmosphere.

Fortunately the equivalent of pure hydrogen can be secured at moderate cost from commercial ammonia which yields when properly processed an extremely pure gas. Verbal description of this gas requires a cumbersome phrase so it is referred to hereafter as ammogas. It contains one part by volume of nitrogen and three parts of hydrogen and is so inactive with carbon that a soft skin does not form on 1.5 per cent carbon steel soaked in it at 1850 degrees Fahr. (1010 degrees Cent.) for 1 hour. Its composition needs no adjustment to the requirements of the steel to be hardened and its

strongly reducing character permits hardening of stainless steel without tarnish.

Although ammogas is practically nondecarburizing there does occur a slow loss of carbon which is made up by diffusion of carbon to the surface from the interior of the steel replenishing that lost from the surface and preventing serious impoverishment of the surface layer. This is accomplished because of the slow rate of carbon transfer across the surface. However when treating very thin sections carbon loss becomes more of a problem due to the lack of sufficient stock to permit proper diffusion of the carbon. When this condition is met the gas may readily be made completely nondecarburizing or actually carburizing by the addition of small proportions of hydrocarbon gas. Thus it is possible to regulate the carbon pressure of the gas between wide limits without decreasing its reducing powers. This is not possible in the high CO type of atmosphere in which the carbon pressure and oxygen pressure are not independent.

Ammogas must be very dry for successful use as a protective atmosphere for tool hardening. A moisture content of 0.03 per cent by volume (dew point -30 degrees Fahr. or -35 degrees Cent.) at the work surface is generally satisfactory, but the drier the gas the better. Contaminating oxygen combines with hydrogen at furnace temperatures to form water vapor. The presence of water vapor can be detected in very small concentrations by a dew-point test. Rate of decarburization increases rapidly with increase in moisture content as shown presently.

It is not sufficient merely to put dry gas into a furnace; it must be maintained dry while it is there in contact with the work. This means that oxygen and oxygen-yielding compounds from the air, from the furnace brickwork, or from any other source must be rigorously excluded. As this strict but essential condition could not be met by the ordinary electric box-type tool hardening furnaces having exposed brickwork it was necessary to build a special furnace designed for use with this atmosphere.

FURNACE STRUCTURE

In designing a furnace to meet the requirements enumerated the intent was to devise a simple means for heating and cooling tools in a pure gas. It was apparent that the heating chamber would require

a gas-tight muffle to keep the protective atmosphere from being contaminated by undesirable gases from the brickwork; therefore a metal muffle was adopted.

To enable moving the work in and out of the heating chamber without admitting air to this zone it was necessary to have a "gas lock" or purging chamber with suitable doors. With such an arrangement the work can be placed in the purging chamber where it remains until the air is flushed out by flow of pure gas. After the purging chamber has been thoroughly flushed the inner door between it and the heating chamber may be opened and the work moved into the heating chamber without letting in air. The reverse procedure is used for removing the work from the furnace.

This purging chamber could be used for cooling the work under the protection of the gas but another chamber was provided for this purpose at the opposite end of the muffle so that the work could be moved through in one direction, entering at one end and leaving at the other. This arrangement increases the capacity of the furnace by allowing different pieces of work to be in various stages of progress through the furnace at the same time.

Fig. 1 is a photograph of the furnace in which these ideas were embodied. It consists of an externally heated metal muffle extending beyond both ends of the furnace body. The extension at the right is the entrance chamber and the one at the left is the water jacketed cooling chamber.

There is a vertically sliding door at each extreme end of the muffle. Also two top-hinged doors are provided inside the muffle to separate the entrance and the cooling chambers from the heating chamber. These inner doors are opened and closed by a lever extending outside the muffle.

The furnace is electrically heated by means of metal resistor straps in sinuous loops. There are two independently controlled heating sections, each serving half the length of the heating zone. The section adjacent to the entrance chamber is a preheat zone and the other section is the high heat zone. The maximum safe operating temperature is limited by the nickel alloy resistors to about 2000 degrees Fahr. (1095 degrees Cent.). By the use of molybdenum resistors or Globar type heaters this limit could be substantially increased to one imposed by the muffle material.

A reducing atmosphere is maintained in the outer shell of the furnace body to prevent oxidation of the heating ribbon and the



Fig. 1—Electric Heat Treating Furnace with an All-Metal Muffle for Hardening Tool Steels Within a Protective Atmosphere of Dissociated Ammonia.

outside surface of the heated section of the metal muffle. A small independent flow of the protective gas is used for this purpose.

Welded nickel plate was used for the heated portion of the muffle, the balance being made of plain hot-rolled steel. Nickel was selected for this purpose because of its high temperature strength and resistance to embrittlement and melting in carburizing atmospheres. The muffle is 5 inches by 10 inches inside dimensions and

11 feet long over all. It is stiffened by arches in the hot zone which limit the thickness of work that can be accommodated to about 3 inches.

The protective gas is fed into the muffle in the heating zone and discharges through adjustable port at the extreme ends of the muffle where it is burned in a flame. The inner doors are loose fitting enough to allow the free outward flow of gas past them and yet they prevent air from getting into the heating chamber during the short period when the end doors are open.

The work is carried through the furnace on a light weight tray shown in Fig. 2. Its inside dimensions are 9 by 15 inches. A 4-mesh heavy nickel wire screen resting on the runners forms the bottom of the tray. This screen can support small pieces and it allows the free transfer of heat to and from the work necessary for hardening heavy pieces. When heavy pieces are treated they are set on the tray so that their weight will be carried directly by one or more of the runners and not by the screen. Long slender parts that might bend readily under their own weight have been successfully hardened by supporting them on a bed of sand carried on the tray on a flat metal plate with turned up edges.

At each end of the tray there are four radiation shields which show in Fig. 2. The purpose of these shields is to aid in attaining a uniform temperature within the tray length by reducing the heat losses out the ends of the muffle. Combined with the shielding effect of the arches, they accomplish this purpose very well. Without them the heating zone would need to be considerably longer to get the same length of uniform temperature. The trays are moved through the furnace by means of a push rod working through a notch in the outer door which remains closed at the time.

APPLICATION OF HARDENING PROCESS

The furnace described here was built primarily for the hardening of blanking and forming dies as well as miscellaneous tools. Practically all of the requirements of these dies or tools can be met by selection from the following types of tool steels, all of which harden by gas cooling even in large sections.

No.	Carbon	Chromium	Molybdenum	Tungsten	Vanadium
1	0.7	4	.	18	1
2	0.8	4	8	1.5	1
3	1.5	11	1	...	0.2
4	1.0	5	1	...	0.2

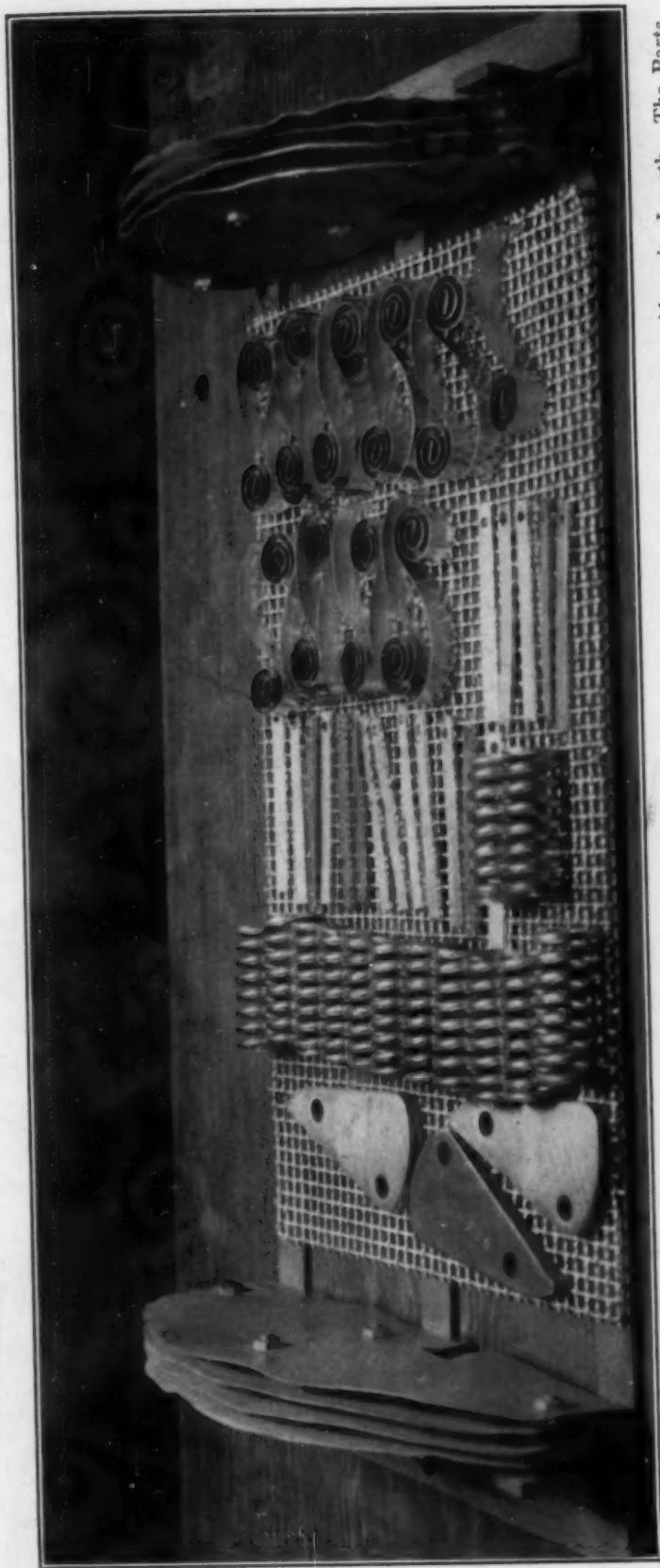


Fig. 2—Light Weight Work Tray with Radiation Shields at Both Ends to Aid in Equalizing the Temperature Along its Length. The Parts on the Tray are 14 Per Cent Chromium Stainless Steels as Hardened by Gas Cooling.

While all have been successfully treated, the present maximum temperature limitation of this furnace excludes hardening of Nos. 1 and 2 to full heat resistance.

A typical application of this method is illustrated by the following description of the hardening of a die block made from the popular air hardening high chromium tool steel, No. 3. The die is machined to finish size with no allowance for clean up after hardening. It is set on a tray and placed in the entrance chamber which is purged in 15 to 20 minutes. The work is then moved into the preheat zone where it is heated to 1300 to 1400 degrees Fahr. (705 to 760 degrees Cent.) and next into the high heat zone where it is soaked at 1830 to 1870 degrees Fahr. (1000 to 1020 degrees Cent.) for one hour or longer if desirable. Since the danger of decarburization is eliminated, soaking times are not dictated by that factor. Next the work is withdrawn into the cooling zone where it is held until cooled to 200 to 300 degrees Fahr. (90 to 150 degrees Cent.) when it may be taken out without tarnish. Caution, however, must be exercised in cooling to room temperature to the extent of avoiding contact between the work and cold liquids or metal surfaces because transformation is still in progress. The total time in the furnace for a particular job runs from 2 to 4 hours depending on the size of the work but of course four separate jobs can be in the furnace at the same time.

As hardened in this manner the die will have a clean bright surface that is as hard as the underlying metal. A uniform hardness of 745 to 770 DPH (63 to 64 Rc) will exist throughout the section, which may be increased to 770 to 795 DPH (64 to 65 Rc) by tempering for 1 hour at 212 degrees Fahr. (100 degrees Cent.).

In the fully hardened state the steel will have a slightly greater volume than before hardening which will decrease with tempering and approach the initial volume as a limit. As these volume changes are characteristic and definite in amount for each steel, their effect on dimensions may be predetermined and compensated for in machining the tool before hardening so that its final dimensions after hardening will be correct.

Fig. 3 shows a blanking die of high carbon-high chromium steel just as it appeared after this hardening treatment. It has a clean bright surface that was fully hard as indicated by the fact that glass could be readily scratched by a corner of the die. Fig. 4 is a photomicrograph of a section through an exposed surface of a specimen

of this same steel, composition No. 3, after hardening and tempering. It illustrates the freedom from surface damage by decarburization that characterizes this method of hardening.

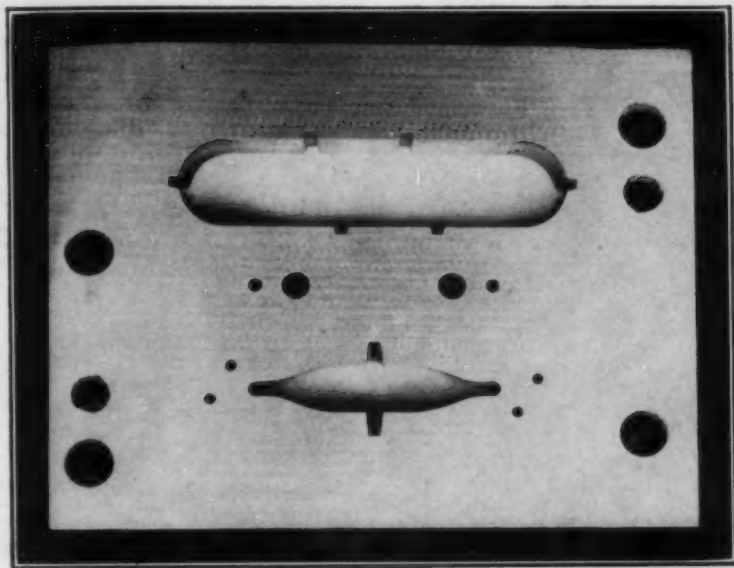


Fig. 3—A Blanking Die of High Carbon-High Chromium Steel as Hardened to 770 DPH (64 Rc) by Gas Cooling. Its Bright Clean Surface Will Scratch Glass and the Need for Finish Grinding or Stoning is Eliminated.

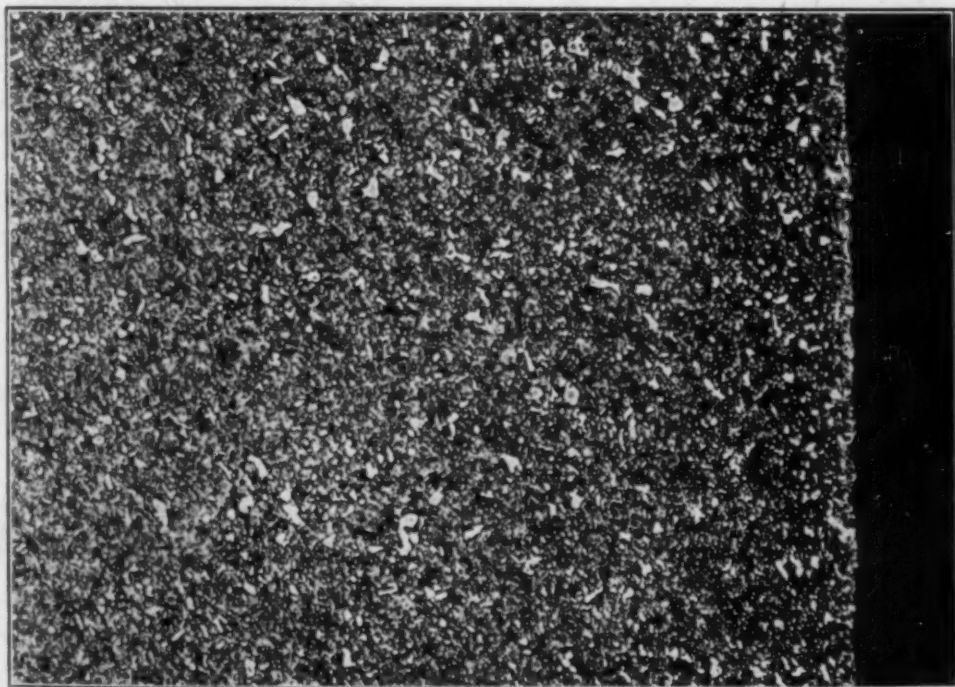


Fig. 4—Section Through Surface of Tempered High Carbon-High Chromium Die Steel After Hardening by Heating for 1 Hour at 1850 Degrees Fahr. (1010 Degrees Cent.) and Cooling in Dissociated Ammonia at Dew Point of -36 Degrees Fahr. (-38 Degrees Cent.). Note Uniformity of Carbides from Interior to Extreme Surface. $\times 150$.

By adding 1 to 2 per cent of natural gas or other hydrocarbon gas to the furnace atmosphere it is made carburizing, although its action is quite slow. In such an atmosphere high speed steels may be given the equivalent of a carburizing pack-hardening treatment. For example a specimen of 18-4-1 high speed steel was heated at 1875 degrees Fahr. (1020 degrees Cent.) for 2 hours in ammogas containing 1.4 per cent methane, then cooled naturally in the gas.

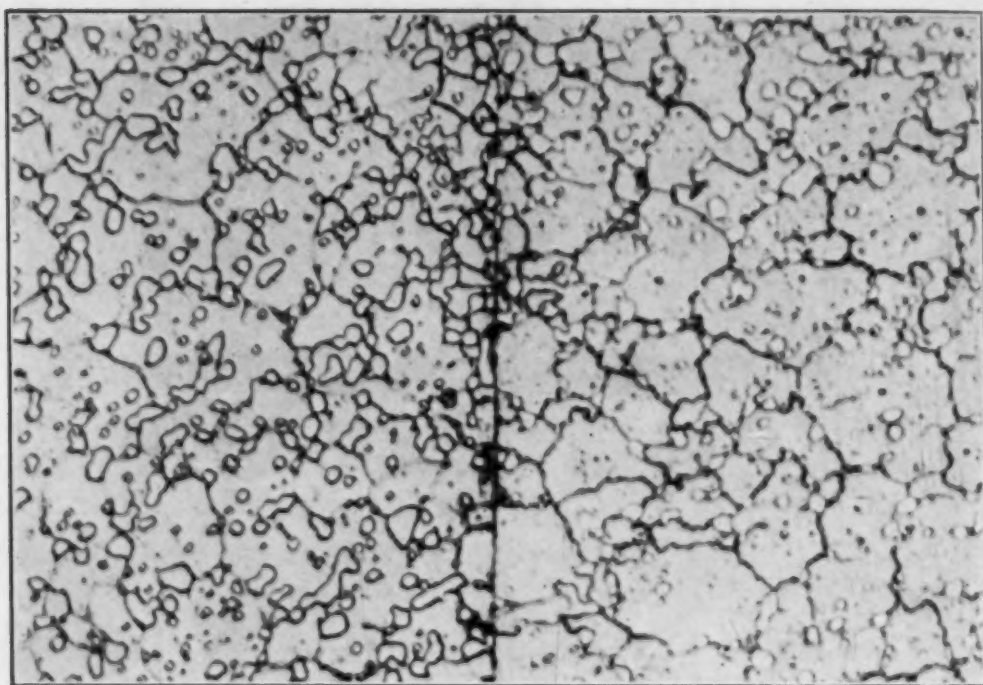


Fig. 5—Section Through Exposed Surface of 18-4-1 and Molybdenum High Speed Steels After Heating for 20 Minutes at 2200 Degrees Fahr. (1204 Degrees Cent.) and Gas Cooling in Dissociated Ammonia. Dew Point —31 Degrees Fahr. (—35 Degrees Cent.). The Center Line is the Interface Between the Surfaces Which Were Clamped Together After Heating. $\times 1000$.

The surface hardness was 880 DPH (67.5 Rc) and the core hardness 645 DPH (58 Rc). This treatment is useful for producing high superficial hardness on high speed steel tools for use where maximum heat resistance is not required.

An experiment was made to determine the possibility of using ammogas for protecting high speed steels at the temperatures normally used for hardening these steels. Specimens of 18-4-1 and of molybdenum-tungsten high speed steels, Nos. 1 and 2 of the composition table, were heated for 20 minutes at 2200 degrees Fahr. (1205 degrees Cent.) in pure ammogas having a dew point of —31 degrees Fahr. (—35 degrees Cent.), and gas cooled. Their hard-

ness was 740 to 795 DPH (64 to 65 Rc) both at the surface and inside. Fig. 5 shows a section through the exposed surface of the two specimens which were clamped together for polishing and photo-

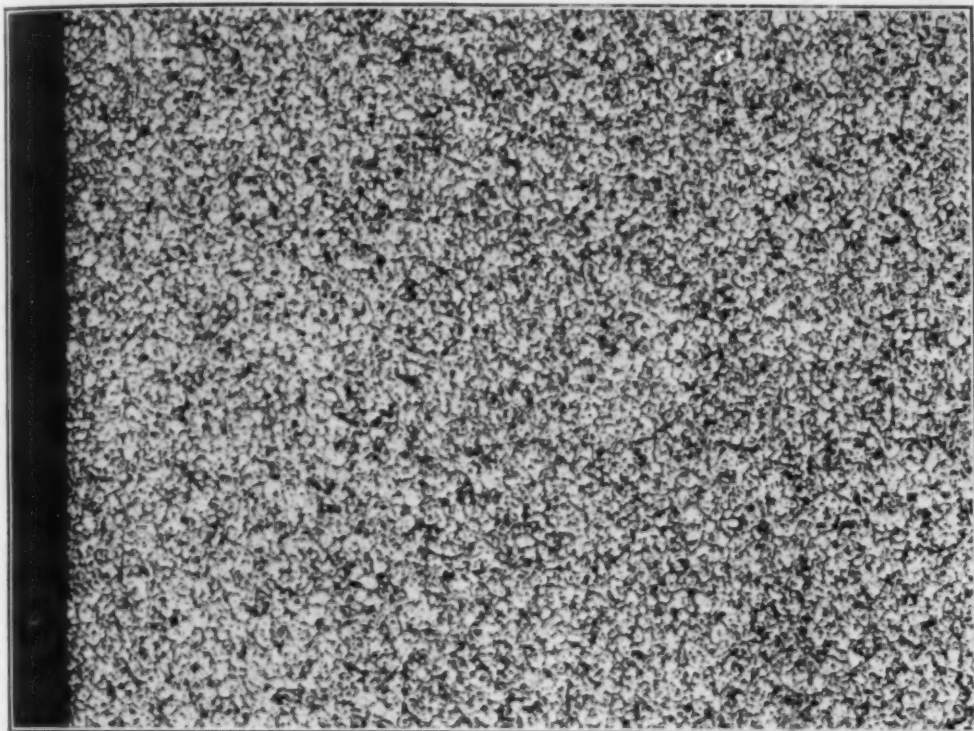


Fig. 6—Surface Section of Molybdenum Steel of Fig. 5. $\times 150$.

graphing; the center line is the interface between them. Fig. 6 shows the high molybdenum specimen of Fig. 5 at lower magnification. There is no evidence of a soft decarburized skin.

Another specimen of 18-4-1 steel was heated at 2250 degrees Fahr. (1230 degrees Cent.) for 20 minutes in ammogas having an intentionally higher dew point of -4 degrees Fahr. (-20 degrees Cent.), and gas cooled. Fig. 7 is a photomicrograph through the surface of this specimen, which was tempered 1 hour at 1000 degrees Fahr. (540 degrees Cent.). The light area indicates a decarburized surface which was confirmed by hardness measurements, the surface being 720 DPH and the core 800 DPH under a 10-kilogram load. The decarburized skin on this specimen was due to the intentionally high moisture content of the gas, proving the necessity of keeping the gas atmosphere thoroughly dry.

Many alloy steels (other than those mentioned here) can be hardened in the same manner when of small section. On the tray

shown in the photograph, Fig. 2, are springs of stainless steel (12 per cent Cr, 0.3 per cent C) as hardened to 630 to 665 DPH (57 to 59 Rc). Their bright untarnished surface is apparent from the re-

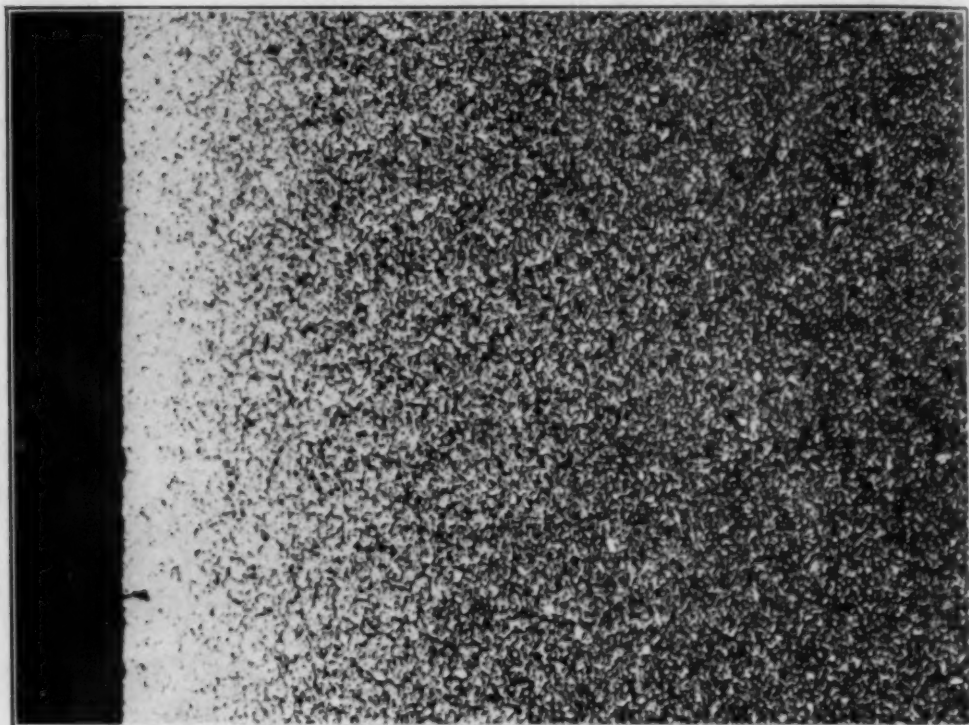


Fig. 7—Section Through Exposed Surface of Tempered 18-4-1 High Speed Hardened by Heating 20 Minutes at 2250 Degrees Fahr. (1232 Degrees Cent.) and Gas Cooling in Dissociated Ammonia of Dew Point -4 Degrees Fahr. (-20 Degrees Cent.). White Zone Indicates Decarburized Surface. $\times 150$.

flections. This is an excellent illustration of the extremely high quality of reducing gas obtained at the work surface. Evidence of its quality with respect to decarburization has already been given.

When this improved hardening method was first applied to dies and other tools for production use its advantages were quickly recognized by Westinghouse tool engineers. These advantages are: freedom from surface damage and distortion which avoids expensive finish operations after hardening, absence of high internal stress which avoids cracking dangers and minimizes the need for tempering, and reliability of the method which gives assurance that each job will be hardened successfully. At the present writing a total of 258 dies have already been hardened for shop production use by this method without a failure or complaint and with many reports of superior performance,

ORAL DISCUSSION

J. B. COOK:¹ I would like to ask what the analysis was, and also the size of the specimen which was heat treated, and the Rockwell hardness after being air-cooled.

J. R. GIER: The high carbon-high chromium steel was 1.5 per cent carbon, 12.00 per cent chromium, and 0.8 per cent molybdenum. The largest solid block that we have hardened in that particular steel was 3 inches thick, 6 inches wide and 12 inches long. That block was heated to 1850 degrees Fahr.; the hardness after gas cooling was 63 Rockwell. On the 5.00 per cent chromium, 1.00 per cent molybdenum, 1.00 per cent carbon type of air hardening steel which was not shown here, we found that a 3-inch block would harden to about 55 Rockwell. However, on grinding it down to 2.25 inches thick and rehardening, the hardness came up to 64 Rockwell. Apparently in that steel somewhere between 2.25 and 3 inches represents the critical thickness for full hardening.

A. H. D'ARCAMBAL:² Will the speaker tell us something about the application of this equipment for the hardening of high speed steel tools, and the type of heating elements recommended for furnaces running up to 2400 degrees Fahr.?

MR. GIER: This type of treatment has only been applied experimentally to the high speed steel tools under very small scale laboratory conditions. We are considering and attempting to devise means of accomplishing this on usefully large scale, but as yet that has not been accomplished.

But we have established that the atmosphere where the purity conditions are maintained as mentioned before is satisfactory for high speed steel as well as for the lower alloy steels. The problem is to develop a furnace in which pure gas can be maintained at the high temperatures involved in treating high speed steels. One factor that may make this problem a little less difficult is the possibility of using lower temperatures than are conventionally used for high speed steel and somewhat longer heating time; this being permissible, we believe, because of the fact that the steels do not suffer damage while being heated.

BRADLEY STOUGHTON:³ I realize that the two authors are on the staff of the Westinghouse Co., but it seems to me it would be desirable if we could have some discussion of whether this sort of thing could be accomplished in a controlled atmosphere gas fired furnace. There are times when a combustion furnace certainly is cheaper and easier to use than an electric furnace, and the question is, could this same type of atmosphere and lack of decarburization be accomplished if we had a combustion type of furnace instead of an electric type of furnace. Perhaps there are some combustion furnace men in the room that would like to take up the cudgel, as it seems to me that this is fundamental. I would like to hear some discussion. Also, I would like to ask Mr. Gier what his experience is in that respect.

MR. GIER: In answer to that question, all I can say is that if you have a muffle that will prevent diffusion of gases through it and will survive the tem-

¹Metallurgist, Kloster Steel Corp., Chicago.

²Manager of sales, Pratt & Whitney Co., Hartford, Conn.

³Professor of metallurgy, Lehigh University, Bethlehem, Pa.

peratures at which high speed steel must be heat treated, then I see no reason why the atmosphere will not apply. In fact I feel confident that it would be satisfactory. But a ceramic muffle is too porous. It will let through the products of combustion of a gas fired furnace, or in the case of an electric furnace it will let through air and defeat the purpose of this atmosphere.

As I mentioned in the earlier part of the paper, to offset unavoidable contamination another approach can be used. This involves the introduction of hydrocarbon gas which can react with the diffusing oxidizing elements. But as soon as that is done the atmosphere becomes active. That is, instead of being comparatively inert as it is in the case of dry hydrogen-nitrogen mixtures, we have to deal with an active gas whose carbon pressure must be balanced. At the high temperatures involved in heat treating high speed steel, if the gas is active the adjustment of composition must be that much closer to the true equilibrium value so that changes in carbon can be avoided. In heat treating miscellaneous low alloy and high speed steels where the stability of the carbides and the carbon content varies over a considerable range the use of such active gas mixtures requires close control of gas composition, whereas for an inert atmosphere the chances of carburization or decarburization through lack of proper adjustment are absent.

To many of you the apparatus and procedure described in this paper may appear to belong in a laboratory more than in a shop. It is true that this furnace was developed in the laboratory but its practical usefulness has been demonstrated over a 2½-year period and we believe that the improved result obtained with it in hardening costly tools and dies amply justifies the use of this method.

SELECTION OF STEEL AND HEAT TREATMENT FOR SPUR GEARS

BY H. B. KNOWLTON AND E. H. SNYDER

Abstract

The selection of the steel and the type of heat treatment for case hardened spur gears is discussed on the basis of the results or service experience and certain laboratory tests. The different types of gear failures are described and discussed. It is shown that for certain types of gears, which must withstand heavy loads for many cycles of stress, the type of failure to be overcome is pitting or spalling of the surface of the gear teeth. Quenching directly from carburizing was found superior to double quenching so far as resistance to pitting was concerned. This was found to be true for a number of different steels.

Some discussion is given of the different methods of testing gears. A special gear testing dynamometer is illustrated and described. A new type of rolling fatigue machine for the determination of resistance to pitting is also described and the results are discussed.

THE subject of selection of steel and heat treatment for gears has been given a great deal of study for some years. Many efforts have been made to devise laboratory tests, which would serve as a criterion for the performance of gears. Before discussing these tests and presenting new data on the subject, it may be well to make some classification of gear steels and methods of heat treatment to be discussed.

In general the gear steels are usually divided into two main classes—First, full hardening steels, which are steels containing about 0.45 to 0.50 per cent carbon with or without alloying elements such as chromium, nickel, molybdenum, manganese and vanadium. These steels may be heat treated by simply heating to the usual hardening temperature and quenching in oil, or in the case of plain carbon steels in water. Quite frequently, however, steels are heated either in

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cyanide or some special bath producing a light case, or are heated in a gas carburizing furnace. This type of treatment produces a very hard case superimposed over a core having approximately the same properties as that produced by heating and quenching without cyaniding or gas carburizing.

The second type includes the case hardening steels. Such steels usually contain about 0.20 per cent carbon, although some work has been done in recent years with carburizing and hardening of steels of approximately 0.30 to 0.35 per cent carbon. The 0.20 per cent carbon steels are usually carburized for a depth of 0.030 inch or more, followed by one or more quenching treatments. When the 0.30 per cent carbon steel is used the depth of case is usually intermediate between that of the cyanide and the ordinary carburized case.

With regard to proper quenching treatments of case hardened steel gears, there seem to be two schools of thought. One favors the use of fine-grained steel and either a single quench directly from carburizing, or cooling in the box followed by reheating to nearly the carburizing temperature and quenching. This produces a structure containing more or less austenite. The other school favors the double quenching treatment; one from a relatively high temperature and the second from a temperature only slightly above the critical range of the case. This produces a fine martensitic structure in the case. Boegehold (1)¹ has shown a higher resistance to bending fatigue for the double quenched martensitic type of case.

Many efforts have been made to devise satisfactory laboratory tests for gear steels and their heat treatments. There is considerable data available on the hardness, tensile and impact properties, which may be obtained by quenching and tempering gear steels. Such data may be of considerable value for gears, which are made of full hardening steel and heat treated without cyaniding or carburizing.

On the other hand, physical properties of low carbon steels can be correlated only with the properties of the cores of case hardened gears. The performance of case hardened gears depends upon the case fully as much, if not more, than it does upon the core. Some efforts have been made to determine the physical properties of the case, but this is very difficult to accomplish. McQuaid and McMullan (2) made transverse bend tests of rectangular specimens, which had a case only on the top and bottom sides.

¹The figures appearing in parentheses refer to the references appended to this paper.

One of the greatest difficulties in devising a satisfactory test for case hardened work is to design a specimen with a case and core, which are proportional not only in size but in stress concentration with that of case hardened gear teeth. Other investigators have attempted to make tests on gears themselves. It is obviously desirable to make these tests as short as possible, but such tests fail to correlate with actual service.

The tooth shear test is a simple laboratory test. The writers have experimented with this for several years, and while it makes a good routine check for parts in production it is not very helpful in determining the merits of new steels or heat treatments. In spite of considerable progress in the design of shear tools, the writer has been unable to produce the same type of break that occurs in service. The same may be said of shock tests made on gear teeth, with the exception that it has been found even more difficult to get duplicate checks on shock tests.

McMullan (3), Almen and Boegehold (4), and others have reported results of tests involving running the gears on a dynamometer. This type of testing comes the nearest to simulating actual service, and probably gives the most reliable results. Boegehold has devised a bending fatigue specimen with a notch producing a concentration of stress similar to that occurring in the gear tooth fillets.

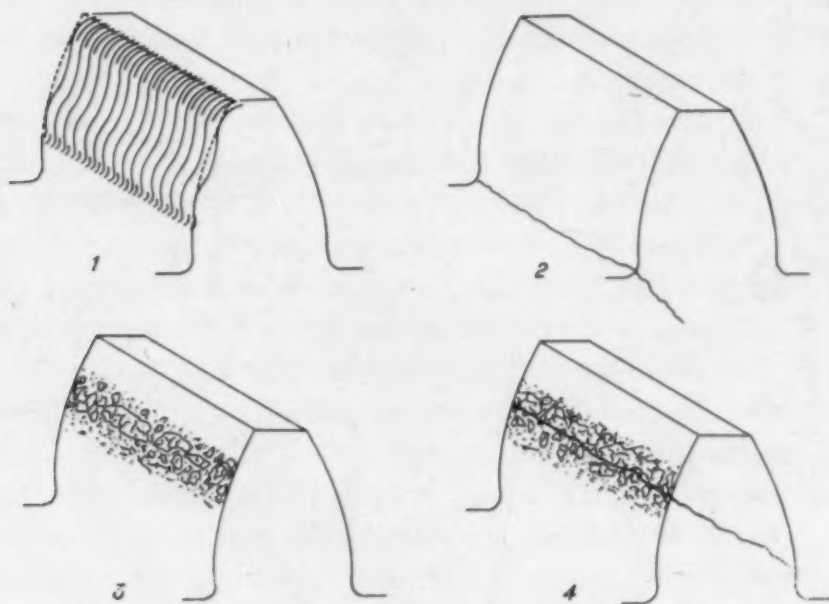
In this paper it will be attempted to discuss gears on the basis of results from both service and dynamometer tests. Also, to discuss briefly the rolling fatigue test, which should have some relation to gear service.

In any study of gear steels and heat treatments, it is necessary to consider the way in which gears fail. The following may be listed as the common types of failure; these are illustrated diagrammatically in Figs. 1, 2, 3 and 4, and photographs, Figs. 5 and 6.

- (a) Wear or abrasion.
- (b) Breaking at the root.
- (c) Pitting of the contact face.
- (d) Breaking through the pitted area.
- (e) Chipping of corners of clashed teeth.

(a) *Wear*—Wear may be due to simple abrasion in which case the worn surface usually becomes smooth. Resistance to wear of this type seems to be directly proportional to hardness. Case hardened gears are more resistant to wear than 0.50 per cent carbon oil hardened gears.

Another type of wear consists of scuffing of the teeth. This produces vertical scratches above and below the pitch line. (See Figs. 1 and 5). This is generally considered to be due to break-down of the oil film and consequent metal to metal contact of the mating



Figs. 1 to 4—The Usual Types of Gear Failures.

Fig. 1—Wear and Scuffing of Tooth.

Fig. 2—Fatigue Break Through the Root of the Tooth.

Fig. 3—Pitting Near the Pitch Line.

Fig. 4—Fatigue Break Through the Tooth Starting from the Pitted Area.

gear teeth. This is encountered particularly with hypoid gears, worms, and other types of gears in which there is a great deal of sliding as well as high pressures. The common method of overcoming this difficulty is the use of an extreme pressure lubricant. These lubricants contain sulphur or some other chemical reagent, which protects the steel by producing an anti-welding surface.

More recently some work has been done with pre-treatments of gears using chemical reagents that will produce an anti-scuffing surface. This surface only lasts during the run-in period, but in some cases it has been found that this is sufficient. If the gears can be run-in with a smooth surface they give no further trouble with scuffing. This cannot be accepted, however, as a universal statement. It probably depends upon the severity of the conditions to be encountered in service.

(b) *Breaking At The Root*—The most common type of failure at the root is illustrated in Fig. 2. This is a bending fatigue type of

failure starting in the fillet at the root and proceeding approximately at right angles to the fillet at the point of origin. In cases where the gears are extremely brittle, it may be possible to break out teeth in some other manner, but the fatigue type of failure is probably the most common.

With case hardened gears this type of failure probably originates in the case. It has been claimed by some that the failure starts be-

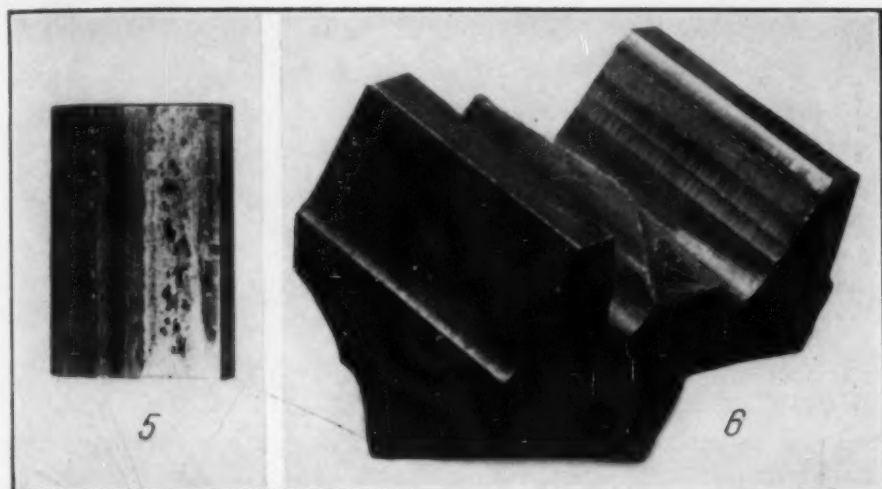


Fig. 5—Scuffed and Pitted Tooth from a Gear Run on the Dynamometer with a Heavy Over-Load.

Fig. 6—Pitted and Broken Tooth from Dynamometer Test; also Over-Loaded.

tween the case and the core. Considering the fact that in bending the stress varies from a maximum at the surface to zero at the center of the core, it seems probable that whether the fracture originates at the surface of the case or in between the case and the core depends upon the relative strength of the case and core and the depth of the case.

In this connection, the concentration of stress at the fillet must also be considered. This also reaches maximum values at the surface; consequently, the stress decreases quite rapidly from surface to center.

The static strength of the case is considerably greater than that of the core. If the case is very thin, the stress in the case portion may be lower than the strength in the case, but the stress at the point where the case and core join may be higher than the strength of the core. In such a case it would be expected that the fracture would originate in the boundary between the case and core. If the case is

of sufficient depth, the stress at the bottom of the case would be less than the strength of the core. Where this condition exists it would be expected that the fracture would start at the surface of the case. If this is the situation the physical properties in the case should be of maximum importance, so far as resistance to bending fatigue is concerned. There is no doubt, however, but that the properties of the core have some effect.

If the gear tooth is hardened all the way through to the same degree of hardness usually obtained in the case, it would undoubtedly be too brittle for many types of service and would fail readily by breakage of the tooth. It is generally accepted that a soft core adds toughness to the case hardened gear teeth. So far as the writers have been able to learn, the exact merits of different physical properties of the core in relation to bending fatigue strength of case hardened gear teeth have never been thoroughly proved.

The work of Boegehold (1) has indicated that the double quenched fully martensitic case has a higher fatigue strength at 250 degrees Fahr. (the usual operating temperature of gears) than does an austenitic case produced by a single direct quench. Gears made of full hardening steel and oil quenched and tempered for a hardness of 50 Rockwell "C" would not be expected to have as high strength at the surface nor as great toughness in the core as case hardened gears.

Another point which cannot be given too much consideration is the effect of the fillet. It is common knowledge that resistance to bending fatigue is inversely proportional to the sharpness of the fillet, or the roughness of the surface. This also applies to gears. The larger the fillet and the more smoothly it is machined, the less will be the concentration of stress or the greater will be the apparent resistance to bending fatigue at the root of the teeth.

(c) *Pitting*—Another type of failure, which occurs with higher stressed gears particularly after long periods of service, is pitting or spalling of the contact face of the gear teeth. This is really a fatigue type of failure similar to that produced by the action of roller bearings, or the action of a car wheel on a railroad rail. One peculiar point is that this type of failure frequently appears after many million cycles of operation.

In bending fatigue testing, it is frequently assumed that any stress which can be withstood for five or ten million cycles of operation can be withstood indefinitely, but this does not seem to be true

with the rolling fatigue produced by the action of the surfaces of two mating gear teeth. Failure probably originates slightly beneath the surface. Frequently a small chip actually flies off the surface of the steel leaving a pitted area.

It is with this type of failure that the present paper will be mostly concerned. Such failures may be due to extreme brittleness of the case where such conditions exist, but they also can be produced by many cycles of operation at high stresses with the best gear steel and best structure produced by heat treatment. It is one of the purposes of this paper to discuss in particular the resistance of various steels and types of heat treated structures to the pitting type of failure.

The depth of the case and the strength of the core have an effect upon the tendency to pitting. If the case is too thin or the core too weak to support the load, severe pitting may result. In this connection, the writer once observed a case where one of two mating gears was badly pitted and the other not. Peculiarly the larger gear was the one which pitted. Microscopic examination showed that the larger gear had a thin martensitic case free from excess cementite. The smaller gear had a relatively heavy case and a more or less complete network of free cementite. Yet it was the larger gear that pitted.

(d) *Breaking Through The Pitted Area*—While this type of failure is not believed to be common in service it has been produced by long dynamometer runs under heavy loads. This is shown in Figs. 4 and 6. This type of failure has also been reported by Ulrich (5). Failure takes place first by very severe pitting in the neighborhood of the pitch line. A fatigue crack then starts in the pitted area and proceeds diagonally downward across the tooth. As stated above, this seems to be uncommon in service. It is mentioned only as a possible evil which might result from pitting.

(e) *Chipping of Clash Gears*—The chipping of the points of teeth of clash gears may be due in part to brittleness of the steel, but tests in the writers' laboratory have shown that design is more important than metallurgy in correcting tooth chipping.

In the present studies two types of tests were employed—dynamometer and rolling fatigue tests.

DYNAMOMETER TESTS

Two specially designed gear testing dynamometers were used in these tests. These are shown in Figs. 7 and 8, and described in detail.

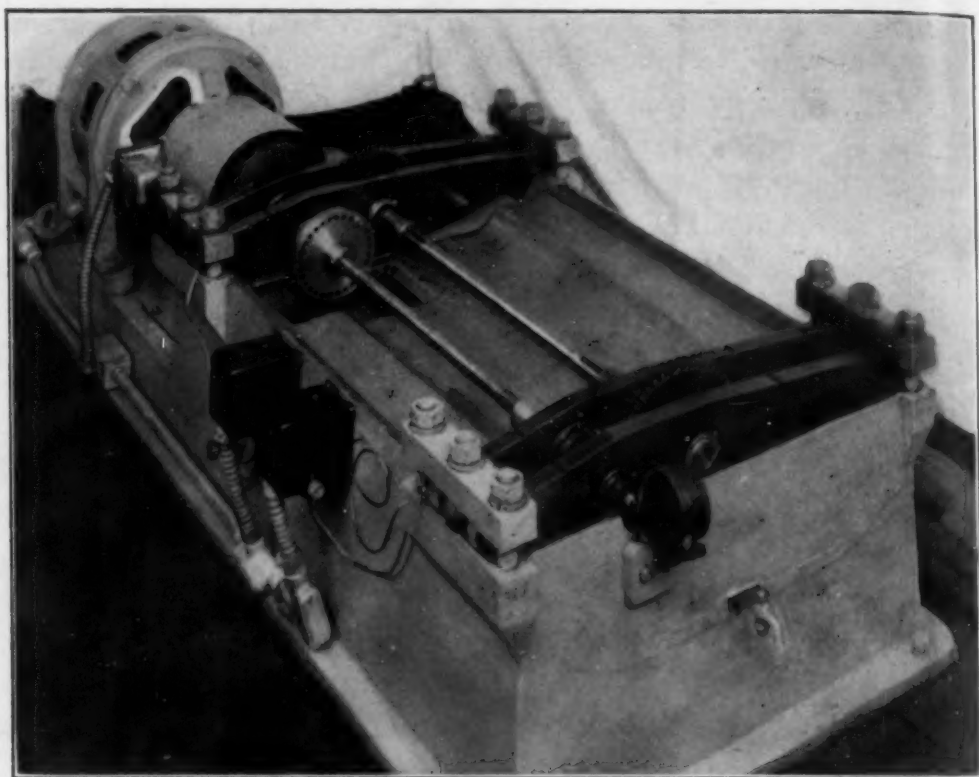


Fig. 7—Gear Testing Dynamometer for Small Gears.

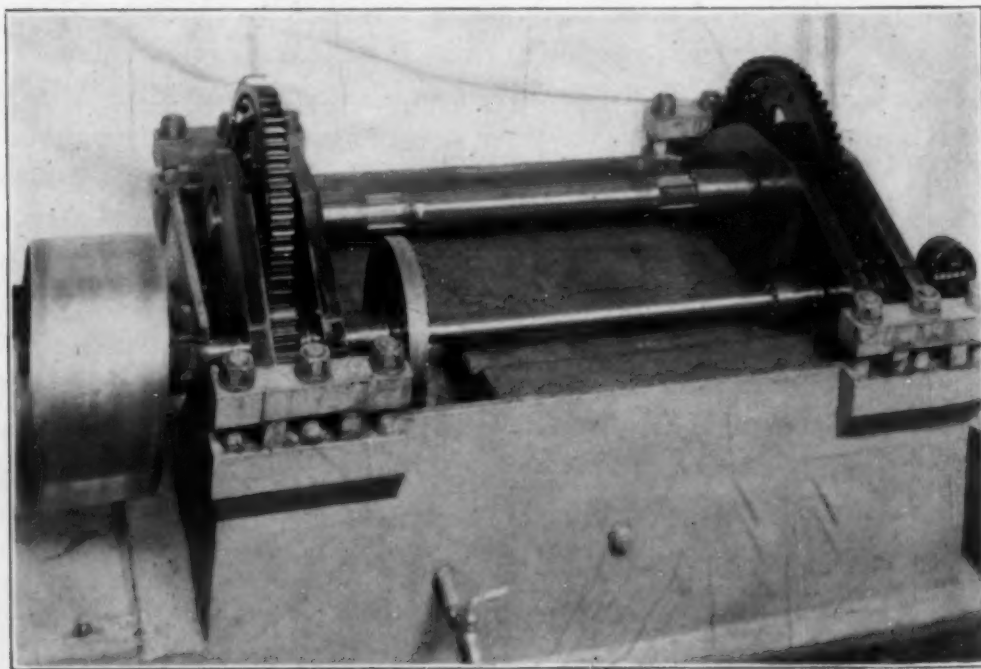


Fig. 8—Gear Testing Dynamometer for 4-Pitch Gears.

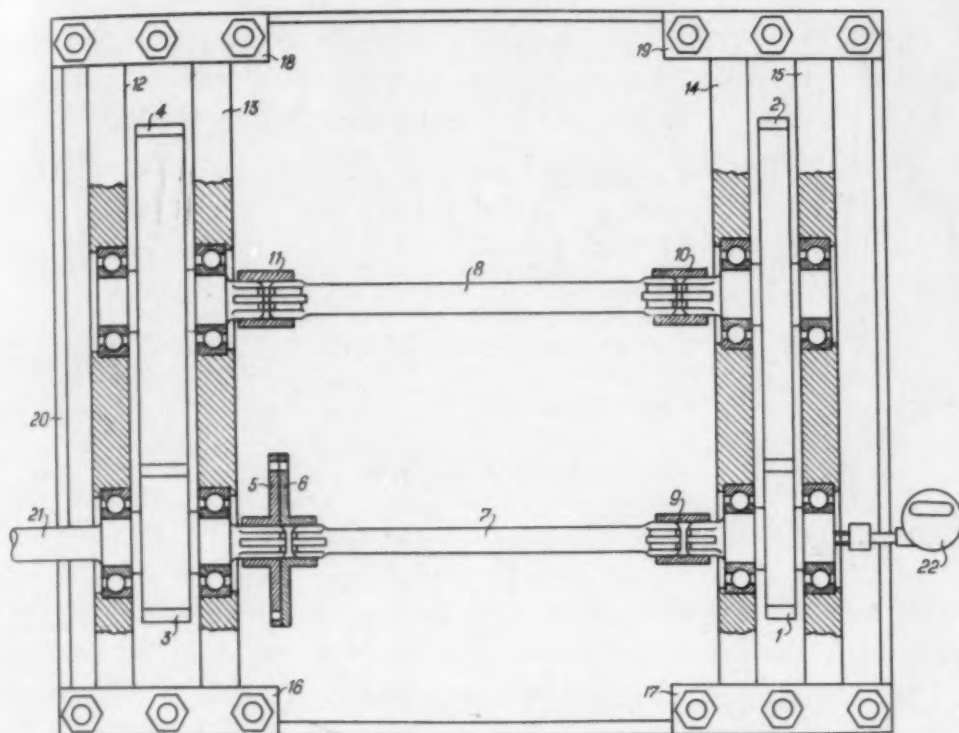


Fig. 9—Diagram of Gear Testing Dynamometers.

With these dynamometers it was possible to maintain a constant torque on the gears without regard to changes in speed or wear of the gears. It was attempted to determine the maximum bending stress and the maximum contact stress, which could be withstood without breakage or without pitting of the surface of the gears. In each case the gears were run for several hundred hours at one load; if this did not produce failure the load was increased for another run. This was continued until some type of failure developed.

TYPES OF STEEL INVESTIGATED

S.A.E. 3145 (modified analysis) and S.A.E. 4145 were used as representative of full hardening steels for 4-pitch gears. Carburized S.A.E. 3115 steel was used in comparison. Six and 7-pitch transmission gears were made of the following case hardening steels:

S.A.E. 4620.

S.A.E. 4120.

S.A.E. 3115.

I.H.C. 1720 (S.A.E. 5120, low chromium).

The following heat treatments were given the case hardening steels:

Carburize 1700 degrees Fahr.
Cool to 1600 degrees Fahr.
Quench directly in Oil.
Draw 380 degrees Fahr.

Carburize 1700 degrees Fahr.
Cool to 1600 degrees Fahr.
Quench in Oil.
Reheat 1480 degrees Fahr.
Quench in Oil.
Draw 380 degrees Fahr.

Carburize 1700 degrees Fahr.
Cool in Pot.
Reheat to 1600 degrees Fahr.
Quench in Oil.
Draw 380 degrees Fahr.

These may be placed in two general groups—one involves a single high quench either direct from carburizing or after reheating and quenching; and the other involves double quenching. The former treatment produces a case with some austenite; while the case produced by the latter treatment is entirely martensitic.

DISCUSSION OF RESULTS

Figs. 13, 14, 15 and 16 show the results of the dynamometer tests. It will be noted that in all cases the single high quench gears are more resistant to pitting than the corresponding double quenched gears. It will also be noted that so far as the type of gears investigated are concerned, pitting occurs before breaking in every case.

With this type of gear it would seem that resistance to pitting is probably of greater importance than high bending fatigue strength. This is further borne out by the fact that pitted gears, which were run for a long time at high stresses, finally broke through the pitted area instead of breaking at the root. This occurred with a full hardening steel gear. However, Ulrich (5) reports fatigue cracks at the pitch line of a case hardened gear. No cracks were found at the root of the same gear.

In this connection, it should be mentioned that the contact stress depends upon the torque, the diameter of the gear, and the width of the face of the tooth, but is independent of the pitch of the tooth. Bending stress, on the other hand, depends upon the pitch of the tooth as well as the other factors. It is possible; therefore, that with the

gears mentioned in this report, the contact stress may be of the most importance, while with gears of other designs it may be that the bending stress is of greatest importance.

While a number of different types of steel were investigated, the number of tests made on each type of steel up to the present time is not sufficient for the writer to wish to give a comparative rating of these steels. Consequently, in the graphs the steels have been blind coded.

It will be noted that in many cases tests were not run at high enough stresses to produce failure by breakage. This is due to the fact that severe pitting was encountered before stresses high enough to break teeth were produced.

The fact that this data shows that the single high quenched gears, at least for the applications under consideration, are superior to double quenched gears may raise a question in the minds of some who made the change from single quenching to double quenching many years ago with apparent improvement in results. There is, however, nothing inconsistent with these two facts.

It is quite probable that the difference between the results obtained 20 years ago and at present are due to differences in steel and in the control of heat treating methods. The steels employed in these tests were fine-grained on the McQuaid-Ehn test; that is, they did not become course-grained at the carburizing temperature. Many of the steels used 20 or 25 years ago became decidedly course-grained during carburizing.

Furthermore, the carburizing furnaces of that period were not as accurately controlled as are the modern furnaces. It is not improbable that occasionally high carburizing temperatures were produced. Under these conditions, gears quenched direct from the carburizing might give considerable trouble, but all of the evidence seems to indicate that with the present steels and present methods of control it is possible to produce superior results with direct quenched gears.

It will also be noted that the tests on the 4-pitch gears showed that the case hardened gears carried a higher load without pitting than did the full hardening steel gears. This does not mean that good gears cannot or are not made from the full hardening steels. Any gear that is designed so that it will not encounter stresses above the endurance limit of the steel used will prove to be a good gear. This has been borne out by the service records on the large gears made of the full hardening steels.

One of the principal objectives of these studies was to determine the maximum safe stress for each type of gear, steel and heat treatment.

Rolling Fatigue Tests—Due to the fact that dynamometer tests consume a great deal of time and require gears which are expensive to manufacture, it has been attempted to devise a laboratory test,

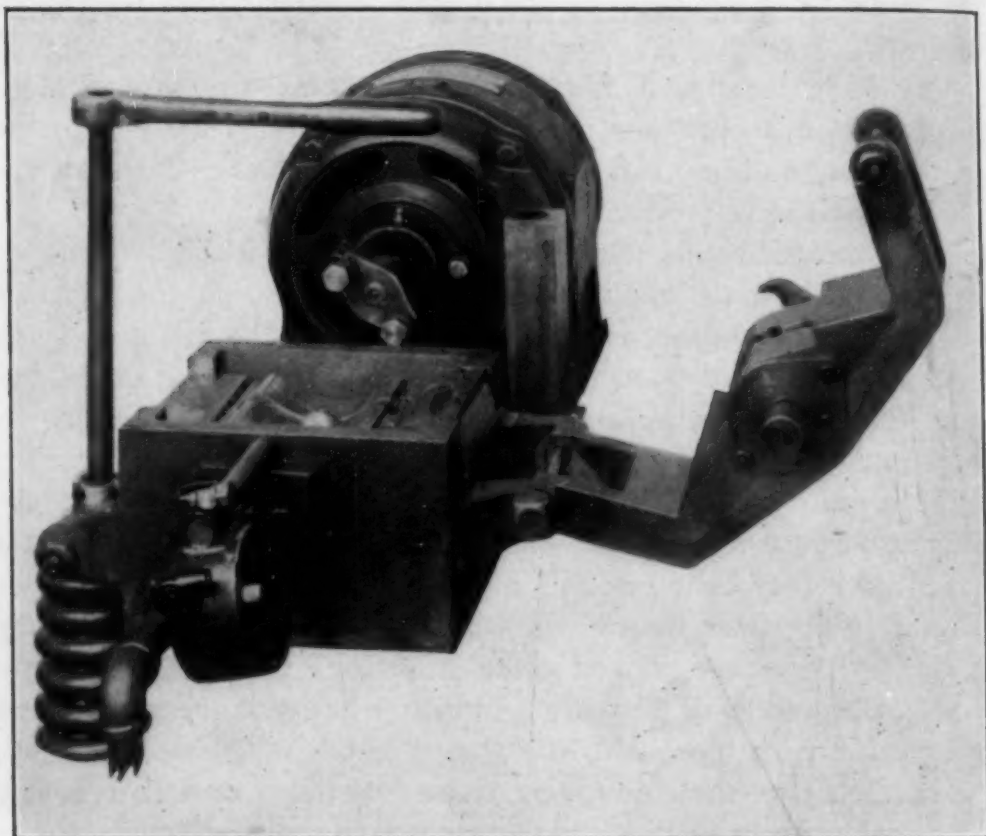


Fig. 10—Rolling Fatigue Machine in Open Position.

which will determine the resistance of different steels and heat treatments to contact stresses. For this purpose a rolling fatigue machine was constructed. Figs. 10 and 11 and the accompanying description show the operation of this machine. This is a modification of the rolling fatigue equipment reported by Way (6).

Briefly, it may be stated that these tests were conducted by running case hardened rollers together under pressure. The radius of the small roller was the same as the radius of the involute curve of a pinion tooth; while the radius of the large roller corresponded to the radius of the involute curve of the gear tooth. The pinion and gear referred to were the ones used in the transmission dynamometer tests.

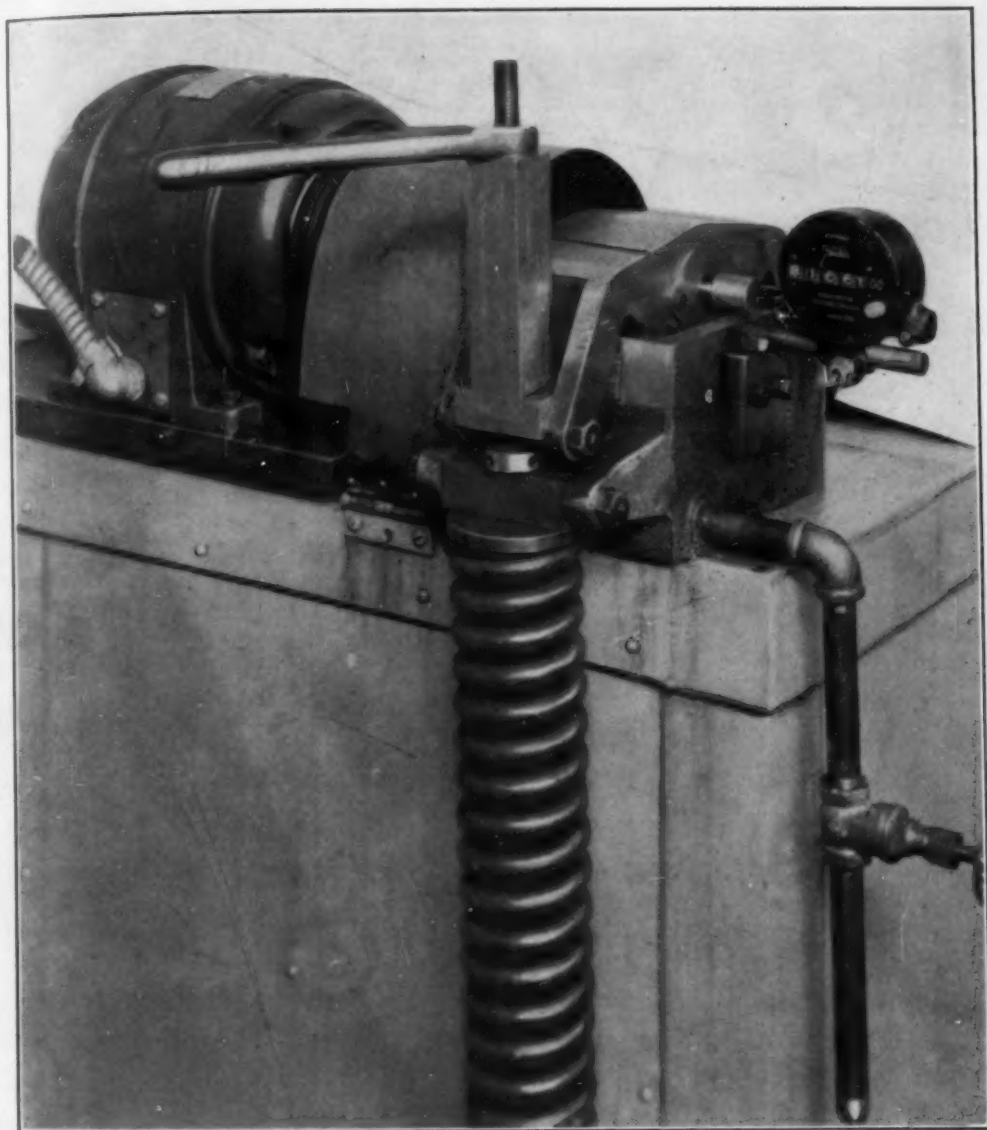


Fig. 11—Rolling Fatigue Machine in Closed Position.

The peculiar thing about the results obtained from these tests is that while the same type of failure was produced on the rollers as the gears, the load carrying capacity of the rollers made of the same steel and given the same case hardening treatment was much higher than that of the gears.

Table I gives a brief recapitulation of the data obtained. It will be noted that all of the steels carried at least 275,000 pounds per square inch and none more than 350,000 pounds per square inch. The type of failure was pitting similar to that produced in the gear tests. The maximum load carried by case hardened gears without

Table 1
Rolling Fatigue Tests

Steel No.	Heat Treatment and Finish	Compressive Strength (PSI)		Beveling	Nature of Failure
		Maximum Carried	Causing Failure		
1	Direct quench, ground after heat treatment	300,000	Not determ.	No	No failure
2	Direct quench, ground after heat treatment	325,000	350,000	Yes	Edges crumbling
	Double quench, ground after heat treatment	325,000	350,000	Yes	Edges crumbling
3	Double quench, not ground after heat treatment	275,000	275,000	Yes	Spalling in decarburized spots
	Direct quench, ground after heat treatment	275,000	300,000	Not determ.	No failure
5	Double quench, ground after heat treatment	275,000	300,000	No	Spalling
	Direct quench, ground after heat treatment	325,000	350,000	Yes	Spalling
52100	Direct quench, not ground after heat treatment	350,000	375,000	Yes	Spalling
	Quench and low draw.				
12% Cr	Ground after heat treatment	300,000	325,000	No	Edges crumbling
	Quench and low draw.				
2% C	Ground after heat treatment	275,000	300,000	No	Edges crumbling

Specimens $\frac{3}{4}$ inch wide were run at 1185 feet per minute and stressed 12,240 times per minute.

Earlier specimens had straight sides; later ones bevelled.

Loads reported as successfully carried were carried for 50,000,000 cycles. Failures were produced after 5,000,000-50,000,000 cycles.

pitting was 184,000-241,000 pounds per square inch; pitting was produced by loads of 198,000-over 241,000 pounds per square inch.

The reason for the higher resistance to pitting of the rollers as compared with gears was not definitely determined. One point of difference in the operation is that the gear teeth are subject to a sliding action above and below the pitch line. This may produce a higher stress in the surface of the metal ahead of the point of contact. It is also possible that distortion during quenching may be less for the roller than for a gear tooth.

It was also thought that the condition of the surface might have an effect upon the resistance to pitting. It was noted that both single and double quenched rollers made of steel No. 2 and ground after hardening carried the same load without pitting, while the roller which was double quenched but not ground failed at a lower load. On the other hand, the direct quenched unground roller made of steel No. 5 carried a higher load than similar roller which was ground after hardening. While the experimental data are still rather meager to draw definite conclusions, this would seem to indicate that slight scaling or decarburizing taking place during reheating for the second quench may lower the pitting resistance. The direct quenched sur-

face seems to be the best. This should naturally be free from scaling or decarburization. Grinding such a surface would not improve it and might even weaken it due to the removal of the highest carbon layer in the case.

Rollers made of S.A.E. 52100 and a 12 per cent chromium, 2.00 per cent carbon tool steel were included in the test for comparison with the case hardened rollers. These showed about the same resistance to pitting as the case hardened rollers with some advantage for the S.A.E. 52100 steel.

Work is being continued on the rolling fatigue test as it seems to give promise of solving pitting problems. If this test can be made to correlate with dynamometer tests there will be considerable advantages due to the speed of the test and the cheapness of the specimens. With the present equipment 50,000,000 cycles of stress can be reached in about 3 days. It may be advisable to design a machine, which will produce both rolling and sliding actions.

Appendix

DESCRIPTION OF SPECIAL EQUIPMENT AND METHODS OF CALCULATING RESULTS

Figs. 7 and 8 are photographs of torsion dynamometers for endurance tests of gears; and Fig. 9 is a diagrammatic illustration of this type of machine, the essential parts of which are as follows:

- 1—Test Pinion.
- 2—Test Gear.
- 3—"Reverse" Pinion.
- 4—"Reverse" Gear.
- 5 and 6—Torsion Disks.
- 7 and 8—Torsion Shafts.
- 9, 10, 11—Coupling Sleeves.
- 12, 13, 14, 15—Bearing Bars.
- 16, 17, 18, 19—Bearing Bar Clamps.
- 20—Base.
- 21—Shaft from Drive Motor or Pulley.
- 22—Revolution Counter.

The torsion disks, 5 and 6, are splined to pinion, 3, and torsion shaft, 7, respectively. In the unloaded condition there is no connection between the two disks except by following around through the train of gears, shafts and other connecting members. If torsion disk, 6, is turned, disk, 5, will also turn but not because of any direct con-

nection between them but because the turning force is transmitted through parts 6, 7, 9, 1, 2, 10, 8, 11, 4 and 3 to 5.

The flexibility of the torsion shafts permits one torsion disk to be rotated through a fraction of a turn while the other is restrained from rotating. The greater this relative turning, the greater the torque that may be imposed on the gear train. By means of appropriate levers and weighing mechanism, this torque may be quite accurately measured. Any desired torque may be retained in the system by bolting the two disks together through a ring of holes

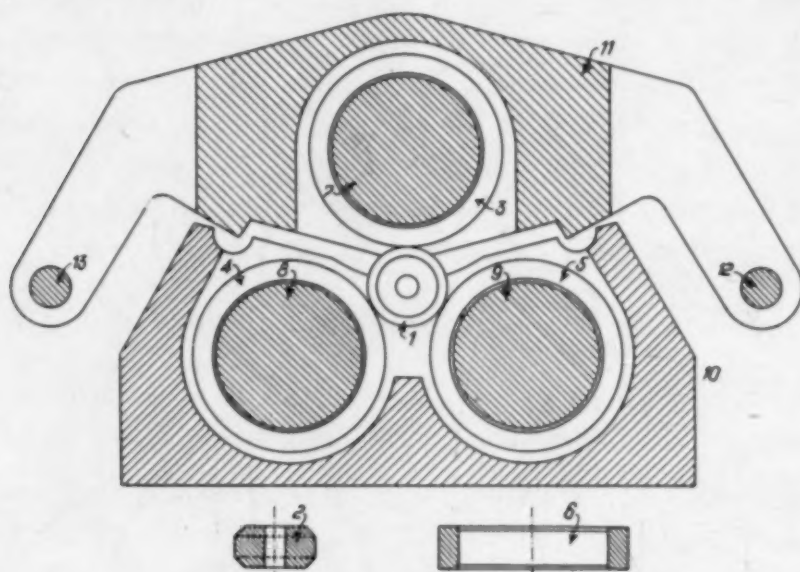


Fig. 12—Rolling Fatigue Machine Diagram.

going around the disks. One of the disks always has one more hole than the other and this results in matching holes being available at very small increments of relative twist. When once set, this imposed torque remains on the gears at all speeds or at no speed at all. There is also a small torque from the required driving force but this is probably always well under 5 per cent of the imposed torque.

Depending on the direction of rotation and the direction of applying the torque, the pinion drives the gear in one end of the machine while the gear drives the pinion in the other. Bending stresses are probably little affected by this, but wear and pitting are quite different and for this reason it is inadvisable to try to run comparable tests in both ends of the machine unless the mating gears have nearly the same number of teeth. Even then there is danger of one set of gears being damaged from failure of the other.

To eliminate such troubles, the "reverse" gears are made wider than the test gears and are permanent parts of the machine. The extra width insures ample strength and a low rate of wear. They are called "reverse" gears because the load is on the reverse side of the teeth from the loaded side of the test gears. They do not rotate in the reverse direction from the test gears.

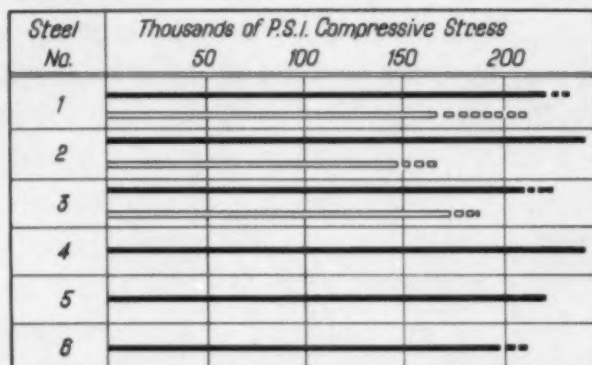


Fig. 13—Compressive Stresses of Carburized Alloy Steel 6-Pitch and 7-Pitch Transmission Gears—1050 Feet Per Minute Testing Speed.

Solid Lines—Stresses Carried Successfully. Dotted Lines—Stresses Causing Failure. Heavy Lines—Single High Quench. Double Lines—Double Quench.

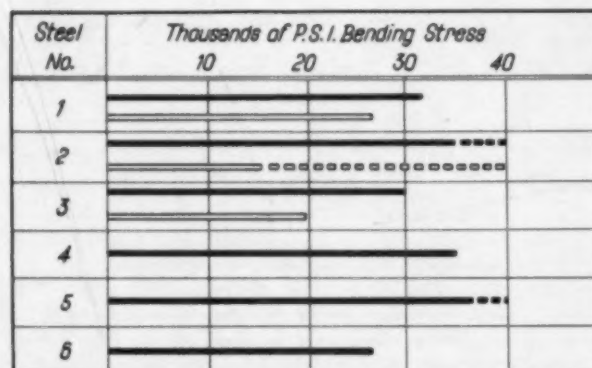


Fig. 14—Bending Stresses of Carburized Alloy Steel 6-Pitch and 7-Pitch Transmission Gears—1050 Feet Per Minute Testing Speed.

Solid Lines—Stresses Carried Successfully. Dotted Lines—Stresses Causing Failure. Heavy Lines—Single High Quench. Double Lines—Double Quench.

Figs. 10 and 11 are photographs of a rolling fatigue machine; and Fig. 12 is a diagrammatic illustration of its essential parts, which are as follows:

- 1—Small Test Roller.
- 2—Small Test Roller, sectional view.
- 3, 4 and 5—Large Test Rollers.
- 6—Large Test Roller, sectional view.

- 7, 8 and 9—Shafts for Rollers 3, 4 and 5 respectively.
 10—Base for Bearings of Shafts 8 and 9.
 11—Lever carrying the Bearings for Shaft 7.
 12—Hinge Pin of Lever 7.
 13—Loading Pin of Lever 7.

Roller, 3, is driven by a direct motor drive to shaft, 7. Rollers, 1, 4 and 5, are driven by friction. Since there is no resistance to rotation except that of the bearings and oil, there is very little slippage. Taking roller, 1, as the center the three large rollers, 3, 4 and 5, are spaced around it at intervals of 120 degrees. With this spacing a normal load put on roller, 3, transmits equal normal loads to both

Steel No.	Thousands of P.S.I. Compressive Stress				
	50	100	150	200	250
2					
7					
8					

Fig. 15—Compressive Stresses of Full Hardening and Carburized 4-Pitch Low Speed Final Drive Gears.

Steel No. 2—Carburized Alloy Steel. Steels No. 7 and 8—Medium Carbon Full Hardening Alloy Steels. Solid Lines—Stresses Carried Successfully. Dotted Lines—Stresses Causing Failure. Heavy Lines—Single High Quench. Double Lines—Double Quench.

Steel No.	Thousands of P.S.I. Bending Stress			
	10	20	30	40
2				
7				
8				

Fig. 16—Bending Stresses of Full Hardening and Carburized 4-Pitch Low Speed Final Drive Gears.

Steel No. 2—Carburized Alloy Steel. Steels No. 7 and 8—Medium Carbon Full Hardening Alloy Steels. Solid Lines—Stresses Carried Successfully. Dotted Lines—Stresses Causing Failure. Heavy Lines—Single High Quench. Double Lines—Double Quench.

rollers, 4 and 5. If roller, 3, has a normal load of 1000 pounds, for instance, rollers, 4 and 5, also each have a normal load of 1000 pounds and not 500 pounds each.

The small test roller, 1 and 2, is beveled. Without such a bevel the edges will usually fail at a lower theoretical compressive stress. The hole through the small roller is for a mandrel during the manu-

facturing operations. It has no shaft or bearings when in the machine. The three large rollers prevent it escaping normal to its surface and it cannot escape in an axial direction because it is in a slot in base, 10; this slot being only 0.001 or 0.002 of an inch wider than the thickness of the roller.

Since the small roller is in contact with three large rollers, it receives three cycles of stress for each revolution. The small roller cannot be much less than $1/6$ the diameter of the large ones; otherwise there is no limit on their relative sizes. For the tests here reported, the large rollers were 2.3 times the diameter of the small one. The small roller was therefore stressed $3 \times 2.3 = 6.9$ times as frequently as any one of the large ones. With roller, 3, being driven nearly 1800 revolutions per minute the small roller was stressed about 17,000,000 times per 24 hours. This is considerably beyond the range of any ordinary gear test and is no doubt only a fraction of the practical speed of this type of test. By reducing the diameter of the small roller to about $1/6$ that of the large ones and increasing the drive speed of roller, 3, to two or three times the 1800 revolutions per minute that has been used, it would be possible to obtain several million cycles of stress per hour.

Lever, 11, hinges about pin, 12, and a load is put on pin, 13, by means of a large calibrated coil spring which is shown in Figs. 10 and 11. This load may be released in a few seconds and the lever swung open as shown in Fig. 10. Small roller, 1, may then be lifted out for inspection. It may be placed under a microscope for more careful examination for damage. Because it is stressed several times as often as the large rollers, it will usually be well on the road to failure before the large rollers show any damage. The large rollers may frequently be used for two or more tests if the small roller is removed before it is too badly damaged.

The two lower rollers dip in oil and throw oil on the upper rollers and bearings. The lever, 11, is made to encourage oil to drain back into the oil reservoir.

The oldest and possibly most commonly used formula for calculating bending stresses in gear teeth is the Lewis formula, which is frequently used in the form

$$S = \frac{P}{F \times P_c \times Y}$$

in which S = Bending Stress in Pounds Per Square Inch.

P = Tangential Pitch Line Tooth Load in Pounds.

F = Tooth Face Width in Inches.

P_c = Circular Pitch in Inches.

Y = Tooth Form Factor.

This formula assumes the tooth carries the entire torque at the tip of one tooth. With well designed gear trains, an effort is always made to avoid this condition and it is much more common for an adjacent tooth to come into mesh at the bottom of the contact area while the first tooth is still carrying the load considerably below the

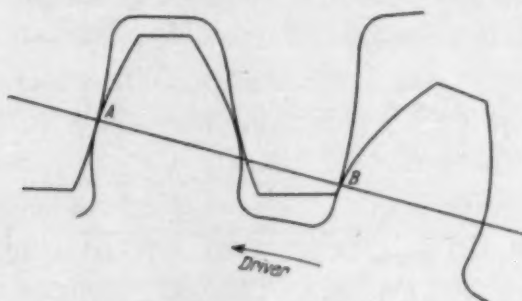


Fig. 17—Method of Determining "Z" Factor for I.H.C. Formula.

tip. The result of this is that the Lewis formula indicates a higher stress than actually exists.

The formula used for calculating the bending stresses here reported is

$$S = \frac{P \times D.P.}{F \times Z}$$

in which S = Bending Stress in Pounds Per Square Inch.

P = Tangential Pitch Line Tooth Load in Pounds.

$D.P.$ = Diametral Pitch in Teeth per Inch of Diameter.

F = Tooth Face Width in Inches.

Z = "Z" Factor.

The "Z" factor serves the same function as the "Y" factor of the Lewis formula but it involves the following factors, which are mostly if not entirely neglected in calculating "Y" factors:

First, it assumes that one tooth carries the entire load not at the point of the tooth but at the point of contact at the instant when the adjacent tooth starts taking up the load. In Fig. 17 this is at point "A"; the next tooth is starting to take up the load at "B". "B" would be at the corner of the mating tooth if there were no tip relief.

Second; back lash, root radii, and undercutting, if any, are considered in determining the section modulus of the tooth, the section modulus being involved in determining both "Y" and "Z" factors.

Since the farthest point at which the tooth carries the entire load is involved in calculating the "Z" factors, it is necessary to know not only the number of teeth of the gear in question, but also the number of teeth in the mating gear.

HERTZ OR COMPRESSIVE STRESS

The Hertz formula in its various forms is probably universally used for compressive stresses between surfaces that are pressed together. The stress varies not only with the total imposed load and with the radii of curvature of the surfaces involved, but also with their moduli of elasticities and Poisson's ratio.

The gear tests here reported were involute steel gears with a 20-degree pressure angle. The formula then becomes—

$$S = 3979.2 \sqrt{\frac{\left(\frac{1}{R} + \frac{1}{r}\right) T}{R W}}$$

In which S = Compressive Stress in Pounds Per Square Inch.

3979.2 = A constant involving steels modulus of elasticity, Poisson's ratio, and the 20-degree pressure angle of the gear.

R = Pitch radius of driving gear in Inches.

r = Pitch radius of driven gear in Inches.

T = Torque of driving gear in Inch Pounds.

W = Effective load carrying width at the Pitch Line in Inches.

For the rolling fatigue tests, the formula used is

$$S = 2266 \sqrt{\frac{P \left(\frac{1}{R} + \frac{1}{r}\right)}{W}}$$

in which S = Compressive Stress in Pounds Per Square Inch.

2266 = A constant involving steels modulus of elasticity and Poisson's ratio.

R and r = Radii of the contacting rollers in Inches.

P = Contact force in pounds.

W = Effective load carrying width of the contact area.

ACKNOWLEDGMENT

The writers wish to express their appreciation of the co-operation of the entire department, particularly Messrs. Sailer, Benton, Halgren, Tinkham and Hodge, in this study of gears. We also wish to give credit to W. E. Henning for the formula for figuring bending stresses in gears.

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DISCUSSION

Written Discussion: By W. P. Eddy, Jr., metallurgical and service engineer, General Motors Truck and Coach, Pontiac, Mich.

The authors should be congratulated for their complete yet concise resume of types of spur gear failures, for their intelligent discussion of heat treating procedures, and for the progress they have made in testing gears and gear steels.

I would like to ask whether or not comparison has been made between the parts double-quenched as described in the paper and parts subjected to double quenching treatment in which the final reheating temperature was approximately 1400 degrees Fahr. I believe it is common practice, with several steels, to quench finally from a temperature in the range 1380 to 1420 degrees Fahr. in order to produce fine martensite cases.

The testing procedures described by the authors should be of considerable interest. We need rapid and inexpensive testing methods which can be correlated with service performance. Certain "wind-up" types of test, notably the four-square test for rear axle gears, have deservedly attained some reputation as valuable methods. The gear test described in the paper, being of this type, would appear to offer considerable promise. The authors' rolling-fatigue test machine is certainly fast, and it is hoped that they will be able to show definite relationship between results obtained with that machine and service performance.

We know too little about computing stress in working parts of all kinds excepting the simplest shapes. Sometimes the lack of realization of how little we really know leads us to place too much reliance on formulas. For ex-

ample, we may report that a certain steel and heat treatment combination, in one type of part or test specimen, withstands a certain stress for a certain number of cycles, but that the same combination, in another type of part, withstands a different stress for the same number of cycles. Would not a sounder conception be that the stressed portion of the steel does not know of what it is a part, and that it would fail at the same stress-cycle combination in any specimen? If so, it then follows that formulas which do not agree are incorrect.

However, it is probably fortunate that we need not, in investigations of this kind, be concerned with actual values of stress. The problem is to establish formulas, each of which can be applied to all parts of a given type. It is of no importance if the stress values computed by such a formula seem fantastic, provided the results of using the formula can be depended upon when applied to actual parts in service. Since we cannot, in most cases, devise suitable formulas with pencil and paper alone, we must obtain them from the results of controlled tests of parts under conditions like those encountered in service. In other words, let us not try to make performance fit formulas, but write formulas to fit performance.

Written Discussion: By E. F. Davis, metallurgist, Warner Gear Division, Borg-Warner Corp., Muncie, Ind.

I have never seen any fatigue failures as indicated by the authors as starting in the pitted area and progressing toward the corner. All I have ever seen have originated at the base of the tooth. The pitch line is not the area of maximum stress, and it seems hardly likely that a fatigue failure would occur through this area.

Otherwise, I concur with everything the authors have said and it is in line with our experience. We have found the direct quenched gear gives longer dynamometer life than the reheated gear, that thin case has produced early pitting, and that chamfer design has a decided effect upon spalling.

One of the most frequent causes of rapid gear wear which the authors did not mention is incomplete meshing of sliding gears, which puts an extreme load upon the tooth and either breaks a tooth or causes the gear to come out of mesh. But this defect, which often is caused by misalignment when mounting the transmission in the chassis, is more mechanical than technical.

A dynamometer test is often misleading as far as determining whether a job will stand up in the field, but is valuable in showing up the weakness of a job such as filleting. The one thing which is most difficult to explain is the wide variance of dynamometer results, which the engineer is most apt to interpret as variables in steel or heat treatment whereas it is most often variables in tooth accuracy, run out, etc.

Oral Discussion

H. F. MOORE:¹ I am inclined to place more confidence in the value of the "rolling fatigue" test than Mr. Knowlton does. In general his results with that test are in harmony with results of "rolling-load" tests on rails carried out in connection with the Rails Investigation now in progress at the Uni-

¹Research professor of engineering materials, University of Illinois, Urbana, Ill.

versity of Illinois. Fissures in shatter-cracked rails were produced only at sections directly under wheel loads. Attempts to produce them by repeated bending moment with maximum moment not under wheel load all failed.

Mr. Knowlton, like all the other students of stress under concentrated loads, goes back to the classical work of Herz. In connection with the failure of rails it seemed necessary to carry Herz' work further, and to determine stresses slightly below the surface. This was done by Thomas and Hoersch,² and while their work was in process of publication it was found that the same mathematical analysis had been worked out by Professor Belaieff and published in Russian. However, the Thomas-Hoersch publication develops an approximate simplified formula, and gives a very ingenious experimental check on the correctness of the mathematical results when applied (statically) to actual metals. The stresses which seem to be the most dangerous under a wheel bearing on a rail, or in the region of contact of one gear tooth on another, are inclined shearing stresses slightly below the surface of contact.

Mr. Knowlton has spoken of the impact on gear teeth. Now the dividing line between static and impact loads is very vague, and up to a very high speed of impact the resistance of the material to fracture seems to be measurable by the energy required to fracture the specimen under static load. Of course, under rather moderate speeds the inertia of structural and machine parts may set up forces and moments not present in static loading, but I am not very greatly impressed with the possibility of there being any great change in the resistance of the material of gears under speeds of application of load common in service.

HARRY W. McQUAID:³ I was interested in what the speaker said in regard to direct quenched versus reheated gears and agree with his statement that the direct quenched gears seemed to give better results than reheated gears. From my own experience I find that carburized gears seem to be definitely better for many applications than lower carbon heat treated gears. The carburized gears, when direct quenched, seemed to be definitely better than reheated carburized gears where the service results in high surface temperature. On automotive truck ring gear and pinion applications, for instance, the pinion is subjected to considerable temperature effect because it is out of the oil bath and cooled and lubricated by spray. The gears, which run in the oil bath, run much cooler and are subjected of course to fewer applications of the load. This results in a high skin temperature on the pinion and for this type of service the direct quenching seems to be advantageous. Apparently the structure obtained by direct quenching is more stable at the operating temperature because of the austenite present. In a dynamometer where temperature effects are, if anything, greater than in service, the direct quenched carburized gear would have a still greater advantage.

G. C. RIEGEL:⁴ I had not had the good fortune to read this excellent paper. There are certain points that seem to me ought to be emphasized in view of the conclusions which Mr. Knowlton stated, and which were further emphasized by Mr. McQuaid, namely, that of the superiority of the directly

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³Assistant chief metallurgist, Republic Steel Corp., Cleveland.

⁴Chief metallurgist, Caterpillar Tractor Co., Peoria, Ill.

quenched carburized gear versus the reheated gear, that is, a gear which has been carburized and reheated for hardening. The only object in making gears is to transmit power and if we are transmitting power on a very limited area of the intended contact surfaces, then the object may be entirely defeated by the distortion which had occurred by quenching from high temperatures. We have found, in our experience, wherein we make gears with a 5-inch face, that we cannot stand for reheating to the high temperature of carburizing for hardening, or for quenching directly from the carburizing temperature without producing coning of the tooth faces which would throw all of the contact surfaces upon a very limited area and therefore defeat the purpose we wanted to obtain.

It is easy to fall into certain conclusions from certain experiments on a small scale and lose out on a larger scale. We have to be careful about our extrapolation of that kind of data.

There is also the other feature on which Mr. Knowlton seemed to lay considerable stress regarding the quality of the steel and in comparing the failures of gears in service. We wish to make this statement, that from our records of gears on which we show permanently the steel identification, that is the mill heat history, the heat treatment history, and the final hardness tests, we are convinced from our own standpoint that not over 3 per cent of the failures could be aided by better selection of steel or better manufacture of the steel at the source. The rest of the improvement must come from increasing the quality of the surfaces in contact, and insisting on making that contact area as generous as the engineers can possibly provide.

To emphasize that phase, we have taken final drive gears, and cut down the tooth width from, say, 3.5-inch face width to as little as 1.5-inch face width, and still carried the load successfully under the most severe condition of service, showing that if we do get perfect contact that much reduced bearing areas can carry the load. It is those misfortunes of average performance in heat treatment and the accompanying distortion which cause the greatest part of our difficulty.

Now, regarding the use of the notched-bar test as a means of determining steel quality, we have made public statement on this before. I think that our results over a long period have borne critical examination and that the notched-bar test does indicate very clearly and very quickly the quality of the mill heat, whether one uses a specimen to test the core properties or combination of core properties and case properties. This method is also a very quick and accurate test of the quality of the heat treatment. Having used this test for quick and accurate determination of these factors, and having correlated such test results with failures in service, we have found that over periods of years where mill heats that were notch sensitive were put into service, that in as many as four or five years in succession, gears bearing those mill heat numbers would come back as failed gears, whereas gears made from mill heats with a clear history of high notch toughness never appeared in subsequent failures. Such a record, made under similar conditions of material, heat treatment, and engineering design, seems to us not to be able to be easily controverted.

Authors' Reply

We were very much gratified with the interest that has been displayed in this paper and the helpful discussions which have been offered. We are particularly interested in Mr. Riegel's discussion. We have always found that his comments are worthy of very thorough consideration.

The writer is in entire agreement with Mr. Riegel concerning the value of mechanical accuracy of gears. Any warping or distortion, which causes a nonuniform distribution of the load along the face of the gears, is likely to be more important than difference in physical properties produced by different heat treatments. It has been our experience, however, that in many cases gears may be quenched directly from carburizing without any more warping or distortion than occurs on reheating and quenching.

With regard to the selection of steel on a notch sensitivity or so-called impact test, we were interested to learn that Mr. Riegel has been able to make some actual correlations between the notch toughness of incoming steel and actual results in the field. We do not doubt that there may be some heats of steel which are inherently brittle enough to lead to failures, particularly with certain types of gears.

However, failures involving breakage of teeth at the roots are due to a combination of notch sensitivity and stress concentration. The smaller and rougher machined the fillet, the higher will be the stress concentration and the greater must be the notch toughness in order to prevent failure of the teeth. On the other hand, if the fillets are made larger and the surfaces machined more smoothly, a lower degree of notch toughness may be entirely satisfactory.

The writer is inclined to believe that more attention should be paid to production of mechanically good gears with large fillets and smooth surfaces, and the metallurgist should be less penalized by being forced to select only those steels which have the highest degree of notch toughness.

While Mr. Riegel did not mention impact tests on gear teeth themselves, this type of test is occasionally employed. The writer does not favor this test as it frequently leads to erroneous conclusions. For example, the impact strength of case hardened gears can be increased by raising the drawing temperature to 600 or 700 degrees Fahr. If the gear teeth can be made soft enough that they will take an actual bend under the blow, the impact strength will undoubtedly be quite high. On the other hand, the fatigue strength is definitely lowered and resistance to pitting and wear is greatly decreased. It must be remembered that gear teeth must not be allowed to take a bend in service; therefore, a gear which is the best on an impact test might not be the best gear from a service standpoint.

Mr. Eddy asks if any "comparison has been made between the parts double-quenched as described in the paper and parts subjected to double quenching treatment in which the final reheating temperature was approximately 1400 degrees Fahr."

No such comparison was made. The second quench was never below 1480 degrees Fahr. The 1400-degree Fahr. second quench, no doubt, gives very high hardness and probably gives gears capable of carrying very high

loads for a few hundred thousand cycles without scuffing or spalling. It may also have a high endurance limit in bending fatigue. For loads which must be carried for many millions—even hundreds of millions—of cycles, however, we believe a quenching temperature as high as 1600 degrees Fahr. is preferable.

Mr. Davis says he has never seen fatigue failures starting in the pitted area and progressing toward the corner. We have seen such failures only three times ourselves, and all three of them were produced in laboratory dynamometer tests. All three of these failures were on low speed final drive gears tested at the low speed under which they would operate in service and carrying over-loads. All of them had the largest root radius that is possible to have between two adjacent teeth.

All three of these gears carried exceptionally high bending loads. They no doubt carried these exceptionally high loads because the full radius between the teeth was less of a stress raiser than the usual small root radius. It therefore appears that these fatigue failures starting from the pitted area are not the result of some exceptional weakness there but instead are the result of the roots of the teeth having been so strengthened that they are no longer the weakest section, especially after some pitting of the contact area has taken place.

Professor Moore says—"I am inclined to place more confidence in the value of the 'rolling fatigue' test than the authors do."

The authors do not have a lack of confidence in the "rolling fatigue" test but believe that as far as gear teeth are concerned, consideration must also be given to the "bending fatigue", which is a major cause of tooth breakage, and to "sliding under load", which is probably the principal cause of scuffing and perhaps somewhat responsible for pitting or spalling.

The rolling fatigue test is very rapid and convenient and when properly run no doubt gives results which are fully as reliable as any other method of fatigue testing, but these results are on rolling fatigue only and do not necessarily bear any direct relationship to the properties of the material in bending fatigue or when sliding under load.

Combination sliding and rolling fatigue tests have been made and probably give a very good measure of the properties required by the contact surfaces of gear teeth. If we try to make specimens and a machine to test them in which rolling, sliding and bending are combined, perhaps we might just as well stick to gears for after all the ultimate product is the ideal test specimen.

Nevertheless, we appreciate Professor Moore's commendation of our efforts to correlate some types of gear performance with rolling fatigue tests. We are continuing our efforts in this direction.

THE INTERRELATION BETWEEN STRESS AND STRAIN IN THE TENSILE TEST

By E. J. JANITZKY AND M. BAEYERTZ

Abstract

Three empirical equations which express the interrelation between stress and strain in the tensile test, as determined on round specimens with a gage length of two inches and a diameter of 0.505 inch, are presented. The validity of these equations is demonstrated by means of experimental data obtained on steels having widely different chemical compositions and various commercial heat treatments such as annealing, normalizing, and quenching followed by tempering.

MANY investigators have examined the effect of individual alloying elements on the tensile properties of steel. With the development of alloy steels and the refinement of heat treating practice, however, the concept of a unique influence on the tensile properties for each element has gradually been abandoned. With proper substitution of alloying elements accompanied by suitable heat treatment, it is almost invariably possible to replace one alloy steel of given tensile properties with other steels which are similar from the standpoint of tensile properties, but may differ greatly in chemical composition. These steels may also differ in other physical characteristics, such as notch impact value and wear resistance.

Under limited conditions of heat treatment, many S.A.E. steels which have the same tensile strength likewise have approximately the same yield strength, total elongation and reduction of area, if the tensile strength is below about 200,000 pounds per square inch.¹ The requisite heat treatment may be any quenching operation followed by tempering, provided the steel and the section be such that the section will harden throughout when quenched. The similarities observed in this restricted group of heat treated steels led to the present attempt to obtain general relations between stress and strain, as customarily determined in the tensile test, which would

¹A.S.M. Handbook, 1939 ed., p. 515-518.

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be applicable to steels in general after any of the commercial heat treating operations such as normalizing, annealing and quenching followed by tempering.

The present study has been confined to tensile properties as determined on round test specimens with a gage length of 2 inches and a diameter of 0.505 inch.

A wide variety of steels was tested. The selection included plain carbon steels, common S.A.E. alloy grades, 12 per cent chro-

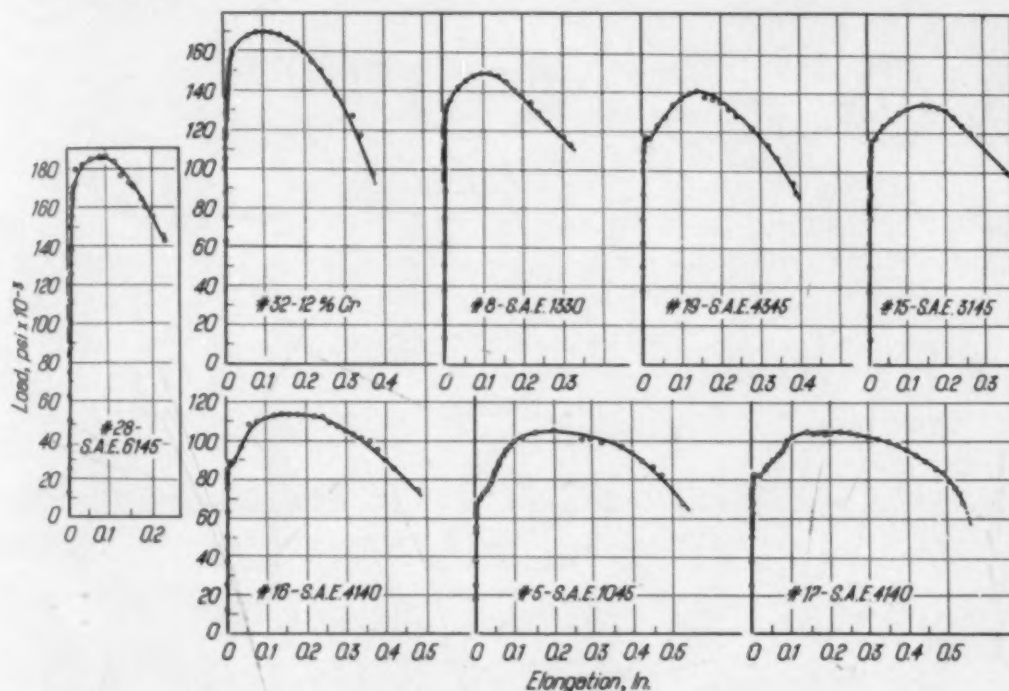


Fig. 1—Load-Elongation Diagrams Obtained with Various Steels After Heat Treatment Given for the Corresponding Number in Table I.

mium steel, and austenitic stainless steel of the 18 per cent chromium-8 per cent nickel type. The heat treatments consisted of quenching followed by tempering, annealing, normalizing, and in the case of the austenitic steel, quenching without tempering. A list of the steels, heat treatments, and the sections in which the heat treatment was applied are given in Table I. In occasional instances duplicate specimens were tested. Duplicates have been grouped together in the table.

After heat treatment, longitudinal tensile specimens of the self aligning type were machined from the centers of bars less than 1.75 inches in diameter, or from a midway position in bars of 1.75 inches and greater diameters. The specimens were pulled in a lever type

Table I

No.	Grade	Section ^a In.	Heat Treat- ment Tempera- tures in °F.	5	6	7	8	9	10	11	12	13	14	15	16	17	18	19	20
				T Psi. x10 ⁻³	Y Psi. x10 ⁻³	B Psi. x10 ⁻³	U Psi. x10 ⁻³	E Psi. x10 ⁻³	R Psi. x10 ⁻³	U (eq. 2)	Dev.	Y (eq. 1a) Psi. x10 ⁻³	Dev.	Y (eq. 2 & 1a) Psi. x10 ⁻³	Dev.	B (eq. 3) Psi. x10 ⁻³	Dev.	B (eq. 2 & 3) Psi. x10 ⁻³	Dev.
1	S.A.E. 1010	1	1650 F.C.†	51.3	26.5	34.0	23.0	40.5	70.4	21.8	-1.2	28.3	+1.8	29.0	+2.5	36.8	+2.8	35.1	+1.1
2	Duplicate			50.0	25.8	34.5	22.5	40.0	71.2	21.2	-1.3	28.0	+2.2	28.6	+2.8	36.2	+1.7	34.3	+0.2
3	S.A.E. 1010	1	1700 A.C.	54.3	30.3	38.0	21.5	37.5	72.5	18.7	-2.8	30.6	+0.3	32.0	+1.9	41.0	+3.0	36.4	+1.6
4	S.A.E. 1010	1	1900 A.C.	54.0	32.3	34.0	21.0	37.5	71.4	18.8	-2.2	30.7	+1.6	32.0	+0.3	39.9	+5.9	36.4	+2.4
5	S.A.E. 1045	1 1/4	1520 W.Q., 1120 A.C.	105.6	69.7	63.0	10.0	27.0	66.5	10.9	+0.9	72.5	+2.8	70.0	+0.3	63.7	+0.7	66.3	+3.3
6	S.A.E. 1050	1 1/4	1650 A.C., 1550 W.Q., 1600 mica	99.0	50.8	83.5	13.5	27.5	46.6	14.7	+1.2	60.6	+9.8	58.6	+7.8	70.1	+13.4	74.8	+8.7
7	S.A.E. 1050	1 1/4	1650 A.C., 1600 mica	99.5	49.5	86.5	14.5	26.5	43.7	14.4	-0.1	59.2	+9.7	59.3	+9.8	78.7	+7.8	78.2	+8.3
8	S.A.E. 1330	1 1/4	1550 W.Q., 900 A.C.	149.3	133.7	111.0	6.0	16.5	56.8	4.9	-1.1	122.2	-11.5	134.2	+0.5	116.7	+5.7	108.5	+2.5
9	Duplicate			149.8	134.0	108.0	6.0	17.5	56.5	5.5	-0.5	122.5	-11.5	127.6	+6.4	109.4	+1.4	106.1	+1.9
10	S.A.E. 2335	1	2100 A.C.	109.1	73.5	97.0	9.0	18.2	38.5	8.1	-0.9	78.2	+4.7	81.9	+8.4	102.2	+5.2	95.5	+1.5
11	S.A.E. 2345	1 1/4	1550 mica	101.0	63.0	78.0	13.8	28.0	50.6	14.3	+0.5	61.0	-2.0	60.1	-2.9	70.4	+7.6	72.2	+5.8
12	S.A.E. 3135	1 1/2	1550 mica	88.0	53.5	64.5	15.5	30.5	59.8	14.8	-0.7	51.9	-1.6	52.9	-0.6	61.0	-3.5	59.1	-5.4
13	S.A.E. 3135	2 1/2	1425, 40° F./hr. to 1200 A.C.	84.5	43.5	66.5	17.5	33.0	55.7	17.9	+0.4	47.8	+4.3	47.4	+3.9	57.9	+8.6	59.1	+7.4
14	S.A.E. 3140	2 1/2	1425, 40° F./hr. to 1200 A.C.	90.8	46.5	73.5	16.3	29.5	45.5	16.9	+0.6	52.3	+5.8	51.6	+5.1	68.7	+4.8	71.0	+2.5
15	S.A.E. 3145	3 1/4	1520 W.Q., 1040 A.C.	134.3	114.4	98.5	7.0	18.0	55.5	5.9	-1.1	104.2	-10.2	112.7	-1.7	104.4	+5.9	97.4	+1.1
16	S.A.E. 4140	1	1550 O.Q., 1250 A.C.	113.8	87.4	71.0	8.0	24.5	66.8	9.0	+1.0	85.4	-2.0	81.1	-6.3	68.8	+2.2	72.0	+1.0
17	S.A.E. 4140	2 1/2	1550 O.Q., 1160 A.C.	104.6	82.0	57.5	9.2	28.0	66.9	11.7	+2.5	74.7	-7.3	67.4	-14.6	58.8	+1.3	65.3	+7.8
18	S.A.E. 4340	4 1/2	1550 O.Q., 1140 A.C.	127.3	97.5	87.8	7.0	21.0	60.1	7.4	+0.4	99.5	+2.0	97.1	-0.4	84.2	+3.6	86.0	+1.8
19	S.A.E. 4345	4 1/2	1550 O.Q., 1140 A.C.	141.1	116.0	86.5	7.0	20.0	63.2	6.7	-0.3	108.5	-7.5	110.4	-5.6	95.8	+9.3	94.2	+7.7
20	S.A.E. 4615	1	1600 F.C.	58.0	36.0	39.0	22.0	43.0	72.3	23.8	+1.8	31.1	-4.9	30.2	-5.8	34.4	+4.6	36.7	+2.3
21	Duplicate			55.6	37.5	36.5	21.0	42.5	73.5	23.1	+2.1	31.5	-6.0	30.3	-7.2	33.6	-2.9	36.2	+0.3
22	S.A.E. 4615	1 1/4	1650 mica	63.0	39.5	38.0	20.5	42.0	67.7	23.9	+3.4	35.1	-4.4	33.4	-6.1	36.8	-1.2	41.8	+3.8
23	S.A.E. 4615	1	1650 A.C.	61.5	41.3	37.0	19.0	41.0	71.4	22.1	+3.1	35.5	-5.8	33.6	-7.7	35.6	-1.4	39.8	+2.8
24	S.A.E. 4615	1 1/2	1700 A.C.	63.0	38.5	42.0	20.0	41.0	70.7	22.3	+1.8	35.1	-3.4	34.2	-4.3	38.1	-3.9	40.8	+1.2
25	S.A.E. 4635	2	1550 mica	87.5	55.2	66.5	11.0	26.0	49.5	12.8	+1.8	59.5	+4.3	55.9	+0.7	60.9	+5.6	66.8	+0.3
26	S.A.E. 6140	1 1/4	1425, 40° F./hr. to 1200 A.C.	74.3	41.7	53.5	17.5	36.0	61.6	19.5	+2.0	42.9	+1.2	41.2	-0.5	46.4	-7.1	50.6	+2.9
27	S.A.E. 6145	1 1/4	1380, 40° F./hr. to 1200 A.C.	85.5	47.7	61.5	17.5	33.8	62.3	17.3	-0.2	48.4	+0.7	48.6	+0.9	59.0	-2.5	55.8	-5.7
28	S.A.E. 6145	1	1700 F.C. to 1500 O.Q., 1000 A.C.	185.9	170.0	143.0	5.0	12.5	39.5	3.9	-1.1	160.0	-10.0	180.0	+10.0	175.0	+32.0	158.4	+15.4

Table I—Continued

1	2	3	4																	
No.	Grade	Section ^a In.	Heat Treatment Tempera- tures in °F.																	
T Psi. x10 ⁻³	Y Psi. x10 ⁻³	B Psi. x10 ⁻³	U E Per Cent			R	U (eq. 2)	Y (eq. 1a) Psi. x10 ⁻³	Y (eq. 2 & 1a) Psi. x10 ⁻³	B (eq. 3) Psi. x10 ⁻³	B (eq. 2 & 3) Psi. x10 ⁻³	Dev.	Dev.	Dev.	Dev.	Dev.				
182.2	167.4	142.0	4.0	12.5	41.3	3.8		-0.2	174.8	7.4	178.2	-10.8	157.4	+15.4	154.8	+12.8				
68.8	34.0	38.0	21.0	42.5	78.4	22.1		+1.1	37.6	3.6	37.0	+3.0	39.3	+1.3	39.9	+1.9				
68.8	34.3	37.5	21.0	42.5	78.5	22.1		+1.1	37.6	3.3	37.0	+2.7	39.3	+1.8	39.9	+2.4				
169.8	143.5	92.0	5.5	19.0	68.3	5.5		+0.0	141.8	-1.7	141.8	-1.7	106.9	+14.9	106.9	+14.9				
168.3	140.2	98.5	5.0	19.5	66.5	5.9		+0.9	147.2	+7.0	136.6	-3.6	102.6	+1.1	103.7	+5.2				
90.6	36.5	51.5	50.0	69.0	78.0	50.5		+0.5	36.5	± 0.0	36.4	-0.1	52.5	+2.1	53.5	+2.0				
88.2	35.3	47.0	58.5	77.0	81.5	58.8		+0.3	34.5	-0.8	34.4	-0.9	52.5	+5.5	53.2	+6.2				
87.5	32.5	58.0	66.0	81.5	80.3	64.7		-1.3	33.5	+1.0	33.6	+1.1	59.4	+1.4	56.0	-2.0				
98.3	37.7	65.5	64.0	79.0	76.3	63.1		-0.9	37.1	-0.6	37.3	-0.4	66.3	+0.8	63.5	-2.0				
35	Duplicate																			
36	18 Cr-8 Ni .05% C	1 ^h	1900 W.Q.																	
37	18 Cr-8 Ni .06% C	1	1900 W.Q.																	

machine of 100,000 pounds capacity, using a speed of 0.11 inch per minute. Up to an elongation of 0.02 inch, the elongation was read from a strain gage. Between an elongation of 0.02 inch and rupture of the specimen, the elongation was measured with calipers.

Load-elongation diagrams were prepared in the usual way by

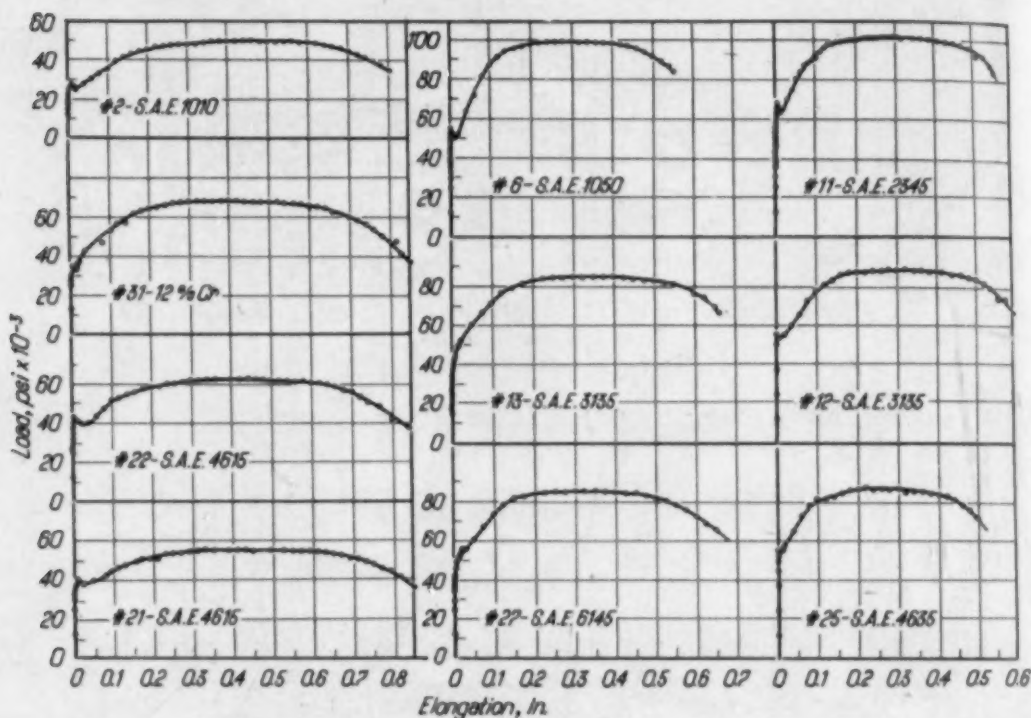


Fig. 2—Load-Elongation Diagrams Obtained with Various Steels After Heat Treatment Given for the Corresponding Number in Table I.

plotting the load, expressed in pounds per square inch of original cross section, against the elongation in inches. Examples of these in Figs. 1, 2 and 3 illustrate the variations in the type of diagram which were obtained with the various steels and heat treatments. Characteristic differences occur in the manner of passing through the yield zone.

For the purposes of this study, the lower yield point was used in instances in which a steel showed an upper and a lower yield point. For the remainder of the steels, the yield strength required to produce a permanent set of 0.2 per cent of the original gage length was used. Hereinafter the words "yield point" will be employed to designate either of these values, depending on the type of diagram obtained with each steel. The uniform elongation was defined as the elongation at the point of maximum load. In instances in which the maximum load remained constant over more than one observa-

tion, the elongation at the center of the flat portion of the load-elongation curve was considered to be the uniform elongation. The tensile strength, the reduction of area, and the total elongation were recorded in the usual way. The breaking strength was determined from the load-elongation diagrams, and usually involved a small

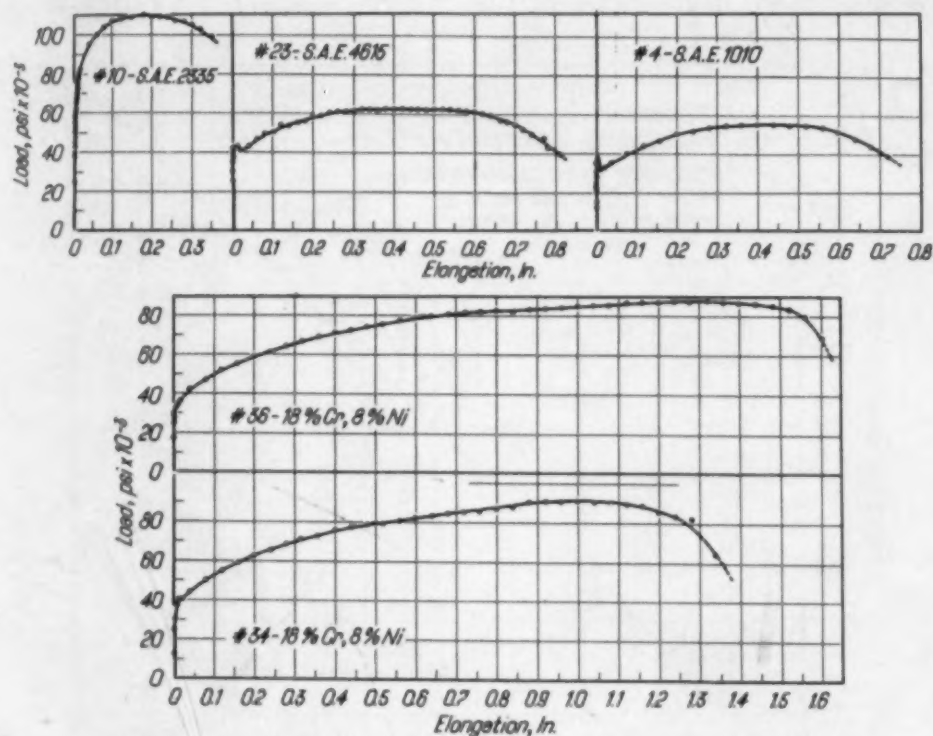


Fig. 3—Load-Elongation Diagrams Obtained with Various Steels After Heat Treatment Given for the Corresponding Number in Table I.

extrapolation of the curve as in Fig. 1, specimen No. 17. The experimental data are listed in columns 5 to 10 of Table I.

Inspection of the data showed that the relation between the yield point (as defined above), Y ; the tensile strength, T ; and the uniform elongation, U , could be approximated by the following empirical equation:

$$\frac{Y}{T^{.84}} = 4.33 \frac{(1+.01U)^{.84}}{U^{.51}} \quad (1)$$

in which T and Y are expressed in thousands of pounds per square inch ($\text{psi} \times 10^{-3}$), and U is expressed in per cent.

Equation (1) may be readily solved for T or Y and tested against the experimental data. Solving for Y ,

$$Y = 4.33 \frac{[T(1+.01U)]^{.84}}{U^{.51}} \quad (1a)$$

Values of Y , calculated by equation (1a) using the experimental values of T and U , appear in column 13 of Table I. Deviations of the calculated from the experimental values of Y are given in column 14. The maximum deviations are $+9.8 \text{ psi.} \times 10^{-3}$ and $-11.5 \text{ psi.} \times 10^{-3}$, and the average deviation is $\pm 4.5 \text{ psi.} \times 10^{-3}$. The deviations are shown graphically in Fig. 4 (black dots). As may be seen

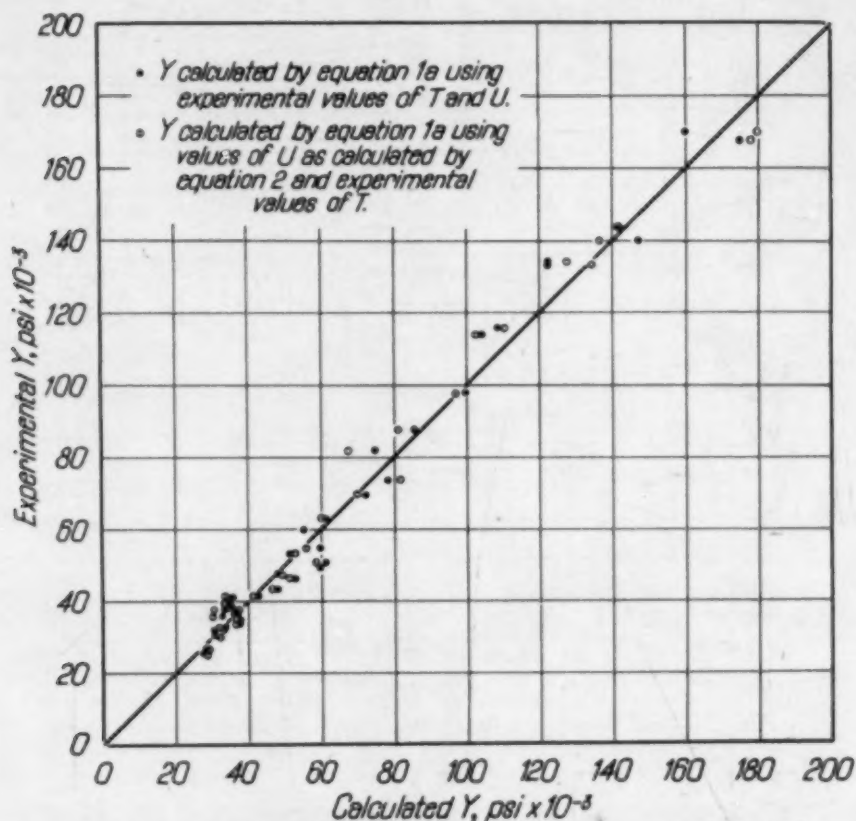


Fig. 4—Diagram Showing Deviation of Calculated Yield Point from Experimentally Determined Yield Point.

from Fig. 4, the deviations in pounds per square inch are as large for low yield points as for high. This makes the possible percentage error larger in the calculation of low yield points.

Since the uniform elongation is not customarily determined in the tensile test, an attempt was made to obtain a relationship between uniform elongation, U ; total elongation, E ; and reduction of area, R . By inspection of the data, the following empirical equation was obtained:

$$U = \frac{E^2}{\sqrt{E^2 + R^2}} + .001 E^2 \quad (2)$$

in which U , E and R are expressed in per cent.

Values of U , calculated by substituting the experimental values of E and R in equation (2), are given in column 11 of Table I. The deviations in per cent U (not percentage deviation) are listed in column 12. The maximum deviations are $+3.4$ per cent and -2.8

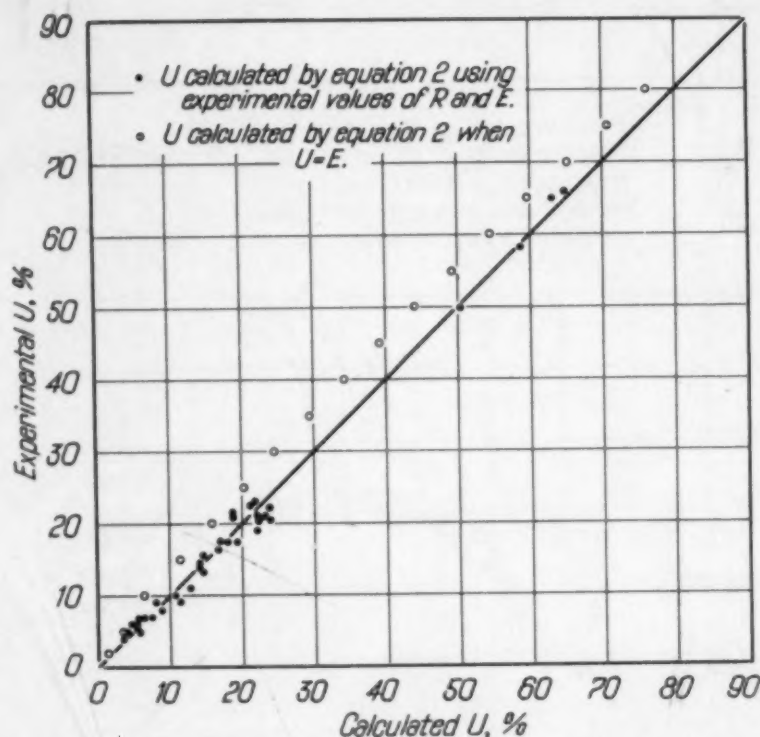


Fig. 5—Diagram Showing Deviation of Calculated Uniform Elongation from Experimentally Determined Uniform Elongation.

per cent and the average deviation is ± 1.2 per cent. The deviations are shown graphically in Fig. 5. The hollow circles in Fig. 5 show the deviation of equation (2) in the limiting case when $U = E$. When $U = E$, the calculated values are lower than they should be, but are sufficiently close to give credence to equation (2) since, so far as we know, this limiting case is not observed in actual steel specimens.

To test the validity of equation (2) for the experimental data, values of Y were calculated by equation (1a), using the values of U as calculated by equation (2) together with the experimental values of T . These values of Y are given in column 15 of Table I, and the deviations in column 16. The maximum deviations are $+10.0 \text{ psi.} \times 10^{-3}$ and $-14.7 \text{ psi.} \times 10^{-3}$, and the average deviation is $\pm 4.0 \text{ psi.} \times 10^{-3}$. The deviations are shown graphically in Fig. 4 (hollow circles). The maximum deviations and the average deviation are

of the same order of magnitude as those obtained when Y was calculated using the experimental values of U.

In addition to the equations expressing the interdependence of T, Y and U, and of U, E and R, inspection of the experimental data showed that the relation between the breaking strength, B, and

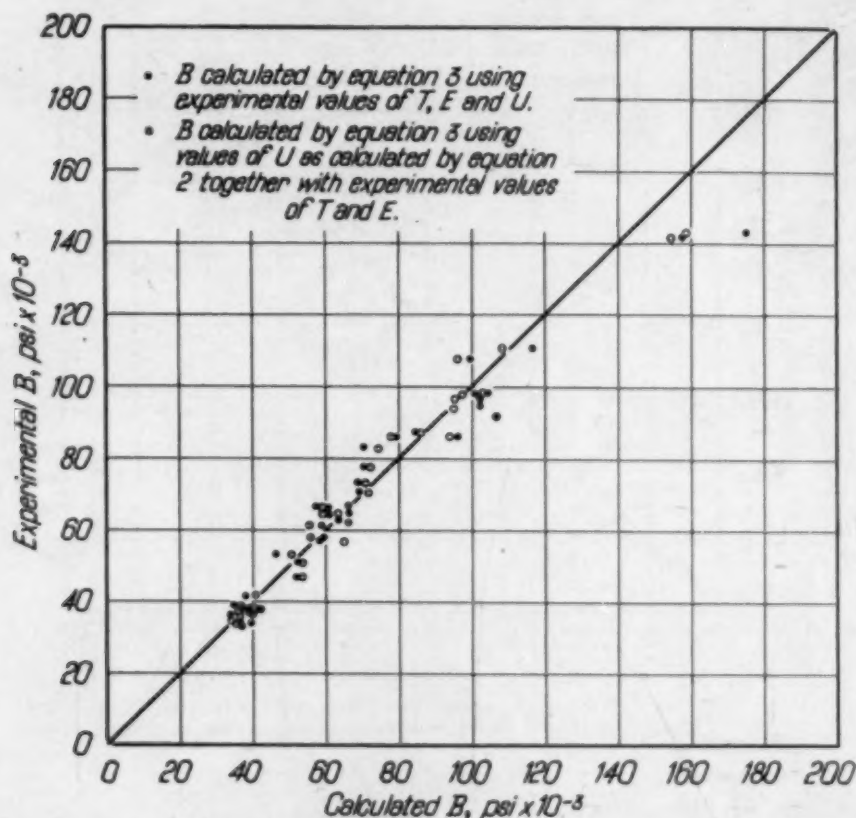


Fig. 6—Diagram Showing Deviation of Calculated Breaking Strength from Experimentally Determined Breaking Strength.

other quantities determined by the tensile test could be approximated by the following empirical equation:

$$B = 16.8 \left(\frac{T}{E - U} \right)^{.73} \quad (3)$$

in which B is the breaking strength, expressed in psi. $\times 10^{-3}$ as calculated on the original cross sectional area of the tensile specimen.

Values of B, calculated by equation (3) using the experimental values of T, E and U, are given in column 17 of Table I. Deviations of the calculated from the experimental values of B are given in column 18. The maximum deviations are $+32.0$ psi. $\times 10^{-3}$ and -13.4 psi. $\times 10^{-3}$, and the average deviation is ± 5.4 psi. $\times 10^{-3}$. The

deviations are shown graphically in Fig. 6 (black dots). It should be noted that the maximum deviation, $+32.0 \text{ psi.} \times 10^{-3}$, occurred with a steel (No. 28, Table I) for which a duplicate specimen (No. 29, Table I) showed a deviation of only $+15.4 \text{ psi.} \times 10^{-3}$. Moreover, it will be shown in the next paragraph that, if the value of U as calculated by equation (2) for specimen No. 28 is substituted in equation (3) instead of the experimental value, the deviation is only $+15.4 \text{ psi.} \times 10^{-3}$.

Values of B calculated by equation (3), using the experimental values of T and E together with the values of U as calculated by equation (2), are given in column 19 of Table I. Deviations of the calculated from the experimental values of B are given in column 20. The maximum deviations are $+15.4 \text{ psi.} \times 10^{-3}$ and $-8.7 \text{ psi.} \times 10^{-3}$, and the average deviation is $\pm 4.2 \text{ psi.} \times 10^{-3}$. The deviations are shown graphically in Fig. 6 (hollow circles).

The relations just presented serve to bring out the great similarity of steels with regard to the properties as determined by the tensile test. It has been shown that steels having very different chemical compositions and various heat treatments fall, with quite good regularity, into a scheme in which T , Y and U vary according to a single empirical equation. It has also been shown that U can be calculated from R and E by a single empirical equation, and that B can be calculated from T , E and U by a single empirical equation.

The authors gratefully acknowledge the co-operation and assistance of many members of the Carnegie-Illinois Steel Corporation. Especial credit is due Messrs. H. P. Langston and P. Henoch for their work in making the tensile tests and in preparing the diagrams, respectively.

DISCUSSION

JOHN M. LESSELLS:¹ The material which is contained in this paper is of some interest in view of the importance to the engineer of ductility in the tensile tests of constructional steels.

Possibly in such an approach to the subject the authors could have profited by consideration of the well-known elongation equation for explaining, at least partly, the mechanics of the actual breakdown. In the light of such knowledge it is clearly seen that an alloy steel deforms more or less locally whereas a soft steel with a larger grain shows a deformation which is more distributed along the gage length.

¹Associate professor of mechanical engineering, Massachusetts Institute of Technology, Cambridge, Mass.

In order that the work of the authors be really of value it should be coordinated with already known knowledge on the subject. The work of Professor MacGregor at the Massachusetts Institute of Technology which has been published² must also be considered.

Viewed in this light it seems that in our present state of knowledge the reduction of area values are of considerable importance in the proper assessing of material values.

R. H. HARRINGTON:³ As we learn more about the metallurgical aspects of the testing, it seems to me that we will learn a great deal more from statistical analysis of tensile data. At this time I would like to ask one question. Do these specific empirical formulae hold regardless of the speed of testing?

F. M. HOWELL:⁴ I do not believe I have very much to say since I have had no opportunity to study the paper. If these formulas are a function of speed of testing, however, I would like to ask, what speed was used in this work?

Authors' Reply

The authors wish to thank those who have added to this paper by their discussion. In reply to Professor Lessells, it was not our intent to controvert the well-known work of Professor MacGregor. Our paper is not concerned with the mechanics of the deformation in the tensile test, but simply presents empirical relations which have been shown to hold for a wide variety of steels and heat treatments, and by which we believe the reasonableness of tensile data or tensile specifications may be checked.

In reply to Dr. Harrington and Mr. Howell, the speed used for testing was 0.11 inches per minute, and it is quite possible that the use of very different speeds would require some changes in the equations.

²C. W. MacGregor, "Relations Between Stress and Reduction in Area for Tensile Tests of Metals," Tech. Pub. 805, American Institute of Mining and Metallurgical Engineers, Feb. 1937.

³Research laboratory, General Electric Co., Schenectady, N. Y.

⁴Aluminum Company of America, research laboratories, New Kensington, Pa.

MACHINING WITH SINGLE POINT TOOLS

BY M. KRONENBERG

Abstract

The increase in production demanded by the National Defense Program necessitates making users of machine tools acquainted with recent developments in metal cutting science, particularly in the field of single point tools.

One of the most important conclusions drawn from recent metal cutting research is the fact that two main relationships exist between the dimensions of chip and the cutting speed. One of these is based on the life of the tool and the other on the capacity of the machine.

In many production shops it is not yet realized that the full utilization of single point machine tools can be achieved only by a proper co-ordination of cutting speed, chip cross section, tool material and the power of the machine; in previous practice the last named factor has often been neglected.

These relationships are developed in the following article, and numerical values and diagrams are given for practical calculation of cutting speeds and cutting forces from simple formulae for various materials of work and tool.

MANY investigations in the field of metal cutting have been carried out since the time of F. W. Taylor (1)¹, resulting in abundant data. These data, however, refer often only to the special conditions of the tests and are not correlated.

The progress made in recent years permits deducing from these tests generalized cutting laws for single point tools, and building up a scientific system from which principles can be drawn for practical application.

Cutting Speed—General. The heat of cutting is injurious to the tool, blunting and finally destroying it. The heat is caused by the overcoming of molecular cohesion in the work material, by friction between tool face and chip and by friction between the work and the flank of the tool.

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

The author is Doctor of Engineering and with the research department of the Cincinnati Milling Machine Co., Cincinnati, Ohio. Manuscript received January 5, 1940.

For practical use in shops and designing offices, cutting speed values have been determined and are included in many engineers' handbooks. The limits of these values are often so wide however, that they do not permit accurate calculation. The wide limits are due to the large number of factors which influence the cutting speed and tool life.

For example, the dimensions of the chip are often not taken into consideration in many tables for the cutting speed. This defect was well known to Taylor, but in spite of his investigations, adequate cutting speed tables are still absent in general practice. The formula which he developed is too complicated for practical application. We need formulae which on the one hand are accurate enough to give reliable values, and which on the other hand can easily be used by engineers. For shop practice, diagrams or instruction sheets can be prepared from such formulae.

The cutting speed is not a constant value even for the same material and tool as assumed in many tables in handbooks. In such tables, a distinction is made between a "roughing cut" and a "finishing cut." These are very inexact concepts. Opinions as to what "roughing" and "finishing" may be, are different in different factories. Production engineers in the heavy industries will often take for "finishing" what will be considered "roughing" by makers of precision machinery. There is no "boundary" between roughing and finishing, owing to the gradual changing of roughing into finishing by the dimensions of chip.

Different investigators have correlated the cutting speed with different variables, viz.:

1. With depth of cut and feed per revolution as separate variables.
2. With chip cross section which is the product of depth of cut and feed per revolution.
3. With tool life.
4. With horsepower developed at the cutting edge.

The first three relationships express conditions which are directly related to one another while the fourth one is related to the fundamental law of mechanics, viz., power equals force times velocity. The fourth relationship is very important but often forgotten. It will be seen later on that this relationship furnishes the key to a better understanding of the complex metal cutting laws. It permits, together with the other relationships, to develop the basic diagram

(Fig. 5) which may be compared as to its importance and limitation with the iron-carbon diagram in the field of metallurgy.

Cutting speed as a function of chip cross section—While many investigators have considered it necessary to treat separately the effects of depth of cut and feed per revolution, such a method when used in conjunction with the limitation introduced by Item 4 above, results

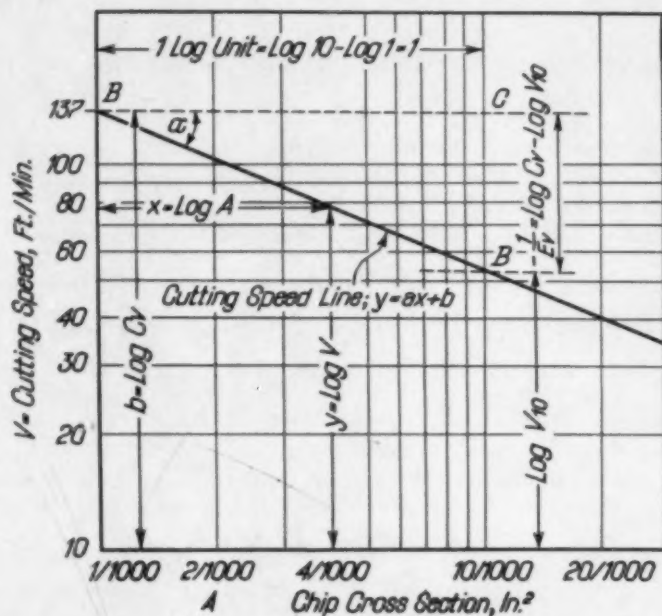


Fig. 1—Derivation of First Cutting Speed Law.

in a very complex procedure. Great simplification is possible by combining these two variables into their product—chip cross section—as tests have proved that this is the main variable upon which the commercially desirable cutting speed depends; provided, of course, that the material to be cut and the tool both remain the same, and that the ratio of feed to depth does not depart greatly from practical values. The bigger the area of chip, the lower must be the cutting speed. The dependence of cutting speeds on chip areas alters for different materials; its variation is greater for the light alloys than for cast iron.

Plotting cutting speed values as a function of the chip cross sections on bilogarithmical papers reveals that this function is a straight line, thus permitting more accurate plotting and easier comparison than would be possible with curves on an arithmetical system. This variation of cutting speed can be seen from the inclination of the line B-B against the x-axis and can be measured by means of the angle α . (Fig. 1).

Introducing the logarithmical values for the variables into the general formula for a straight line

$$y = a x + b$$

leads to the following form of a cutting speed law (5).

$$V = \frac{C_v}{\sqrt{E_v} \sqrt{1000 A}} \quad (1)$$

This may be described as the first cutting speed law or "tool-law," where:

C_v = a constant depending upon material, tool, tool life and cutting fluid.

E_v = exponent depending upon the material to be cut.

A = chip cross section in 0.001 square inch.

v = cutting speed feet per minute for some definite tool life.

Numerical values for C_v and E_v respectively are given in Tables I and II. They have been computed from tests made by different investigators in this country and abroad.

Table I
Values for C_v and E_v of Various Metals for Cutting Speed Law (Formula 1)

Metal to be Cut	C_v For 18-4-2 high speed steel tool*			E_v
	For tool life 60 min. without cutting fluid or 480 min. with cutting fluid	For tool life 60 min. with cutting fluid	For tool life 480 min. without cutting fluid	
Light alloy	2160**	1.37
Brass (80-120 Brinell)	575	1.62
Cast brass	365	2.28
Cast steel	131	182	91	2.75
Carbon steel:				
S.A.E. 1015	258	360	180	2.44
S.A.E. 1025	206	288	144	2.44
S.A.E. 1035	164	230	115	2.44
S.A.E. 1045	131	182	91	2.44
S.A.E. 1060	84	118	59	2.44
Chromium-Nickel Steel	141	198	99	1.77
Cast iron:				
100 Brinell	187	260	130	3.6
150 Brinell	119	168	84	3.6
200 Brinell	67	94	47	3.6

*Correction for other tools see Table II.

**No cutting fluid if turning light metals!

The method of determining C_v and E_v permits moreover of the comparison of the tests from different investigations.

The smaller E_v , the greater is the influence of chip area variation on the cutting speed variation. C_v increases if cutting fluids are used and decreases if a longer tool life is desired. Tool life is understood to be the cutting time of the tool between successive grindings.

Table II
Correction of C_v for Different Compositions of Tools

Type	Approximate Composition Per Cent						Multiply C_v by:
	W	Cr	Va	C	Co	Mo	
14-4-1	14	4	1	0.6-0.8	0.83
18-4-1	18	4	1	0.7-0.75	0.94
18-4-2	18	4	2	0.8-0.85	..	0.75	1.0
18-4-3	18	4	3	0.85-1.1	1.08
18-4-1 + 5% Co	18	4	1	0.7-0.75	5	0.5	1.11
18-4-2 + 10% Co	18	4	2	0.8-0.85	10	0.75	1.28
20-4-2 + 18% Co	20	4	2	0.8-0.85	18	1.0	1.33
Cemented carbide tools (on steel) up to							5.0
Formula for determination of C_v if a tool life of "T" min. is required instead of "60" min.:							
$C_T = C_{v60} \cdot \sqrt[6]{\frac{60}{T}}$							

The use of the cutting speed formula 1 will be clear from the following example.

Required: Cutting speed for turning S.A.E. 1035 with a high speed tool 18-4-2, the area of the chip being $A = 0.013$ square inch, a cutting fluid being used, and the desired tool life being $T = 60$ minutes.

Procedure: Table I: $C_v = 230$

$$E_v = 2.44$$

$$\text{Therefore: } V = \frac{230}{\frac{2.44}{\sqrt{13}}} = 80 \text{ ft./min.}$$

This equation can be solved with the aid of a bilogarithmic slide rule, or—as shown later on—graphs can be prepared which permit of a direct reading of the cutting speed for different materials, tools and chip cross sections.

Cutting speed as a function of tool life—Taylor showed that the relation between the life of a high speed steel lathe tool and the cutting speed could be represented approximately by the empirical equation

$$VT^n = C_T \quad (2)$$

in which V = cutting speed (feet per minute).

T = tool life (minutes).

C_T = cutting speed constant corresponding to 1 minute tool life; it is dependent for its value on the exact cutting conditions other than speed. It will vary with the composition, form, size and treatment of the tool, the material cut, the size and shape of cut, and the cutting fluid.

n = constant representing the variation in tool life with cutting speeds the value of which also depends on factors such as tool material, material cut, size and shape of cut, and cutting fluid.

It is convenient to represent the cutting speed—tool life relationship graphically on logarithmic co-ordinates, whereupon it takes the shape of a straight line of slope n .

This empirical relationship between tool life and cutting speed

in turning was confirmed by the Lathe Tool Research Committee (2) with high speed tools and by the National Bureau of Standards for cutting $3\frac{1}{2}$ per cent nickel steel with high speed steel tools for both roughing (4) and shallow cuts (7) and with cemented tungsten carbide tools for roughing cuts (8) and by many others (6), (10), (14).

Table III
Numerical Values for the Exponent "n" in Cutting Speed Formula (2)
 $VT^n = C_T$

Author	Material	Tool	n
Taylor	Plain carbon steel 70,000 tensile strength	19 per cent C, 2 per cent Cr, 8.5 per cent W	1/8
Digges	$3\frac{1}{2}$ per cent nickel steel	Cemented carbide	1/5
Digges	$3\frac{1}{2}$ per cent nickel rough cut	High speed steel	1/7
Digges	$3\frac{1}{2}$ per cent nickel shallow cut, dry	High speed steel	1/10
Digges	$3\frac{1}{2}$ per cent nickel shallow cut, land oil	High speed steel	1/8.7
Digges	$3\frac{1}{2}$ per cent nickel shallow cut, water	High speed steel	1/7.6
Boston	S.A.E. 3140 dry	High speed steel	Average 1/7.8
Boston	S.A.E. 3140 lubricant	High speed steel	Average 1/7.8
Schlesinger	Corresp. to S.A.E. 3325	High speed steel	1/5.4
Schlesinger	Corresp. to S.A.E. 1045	High speed steel	1/6.6
Judkins and Hecker	S.A.E. 1040 dry	Cemented carbide	1/6.4
Judkins and Hecker	S.A.E. 1060 dry	Cemented carbide	1/6.0
Judkins and Hecker	S.A.E. 2335 dry	Cemented carbide	1/6.2
Judkins and Hecker	High carbon—High Chromium steel	Cemented carbide	1/6.4

Values for the exponent n are given in Table III. It will be seen, that "n" varies with different materials, although no general rule can be found either as far as the tool, the material, or the cut is concerned. This may be due to differences in determining the instant at which the tool is considered to be dull, and also the changes in type of chip flow under changes in cutting conditions (17). A scientific method of determining the instant of tool dulling was discovered in 1913 by Schlesinger (15), viz., the sudden increase of the horizontal components of the cutting pressure. For practical purposes the value n may safely be assumed as $\frac{1}{6}$, due to the fact that a variation of n from $\frac{1}{5}$ to $\frac{1}{10}$ has only a slight effect on the result.

The formula for C_v for tool life other than 60 minutes, (as given in Table II) is derived from formula 2.

Fig. 2 has been prepared for an easy determination of the change in cutting speed if the tool life shall be changed and vice versa.

Example: Required the necessary cutting speed for 1000 minutes tool life in the case that a cutting speed of 80 feet per minute gives an endurance of 60 minutes.

Procedure: Trace from point "80" feet per minute in the scale up to the endurance line "60 minutes" (arrow a) go horizontally to the tool-life-line for 1000 minutes (arrow b) and read the result (arrow c) in the scale; 50 feet per minute.

Cutting speed as a function of depth of cut and feed per revolution.—Formulae for the cutting speed in which the feed and depth of cut are given separately are usually presented (4) in this form:

$$v = C \cdot d^{-x} f^{-y}$$

which can be transformed into

$$v = \frac{C}{A^{\frac{1}{2}(y+x)} \cdot S^{\frac{1}{2}(y-x)}} \quad (3)$$

where:

A = area of cut

S = shape factor $\frac{f}{d}$

d = depth of cut

f = feed per revolution

y, x = exponents

this formula being valid for 60 minute tool life if the constant C is determined accordingly.

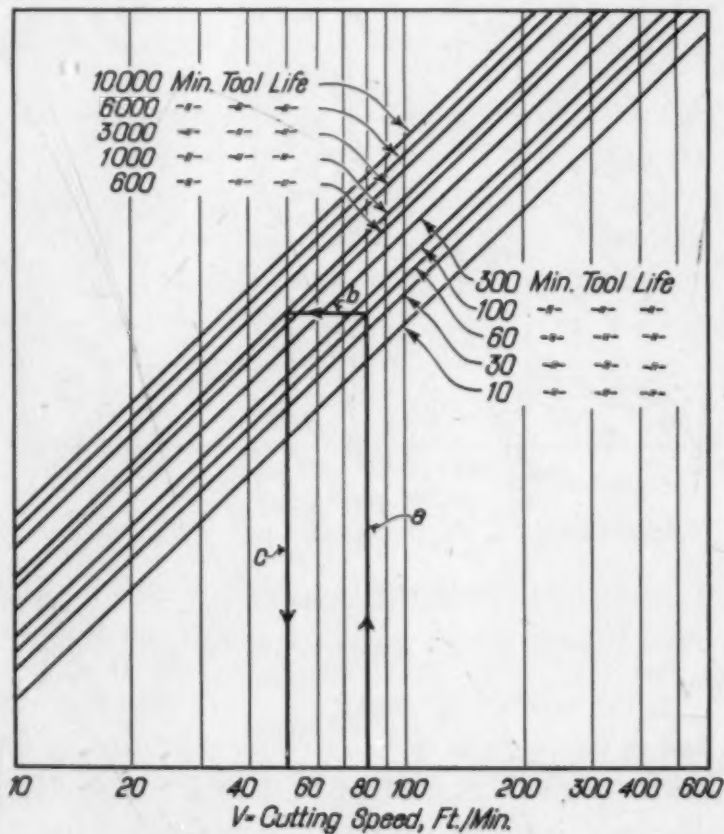


Fig. 2—Effect of Change in Cutting Speed on Tool Life.

This transformation indicates very clearly that the area A of the chip is of a considerably greater effect on the cutting speed than the

shape S , since the exponent of A is composed of the *sum* of y and x , while the exponent of S derives from the difference $y - x$.

Formula 3 shows furthermore, that the cutting speed line (Fig. 1) can be split up into several parallel lines as shown in Fig. 6, for different values of the shape factor S , which in actual practice—in turning—varies approximately between the limits of $\frac{1}{2}$ and $\frac{1}{20}$, and usually only between $\frac{1}{5}$ and $\frac{1}{10}$.

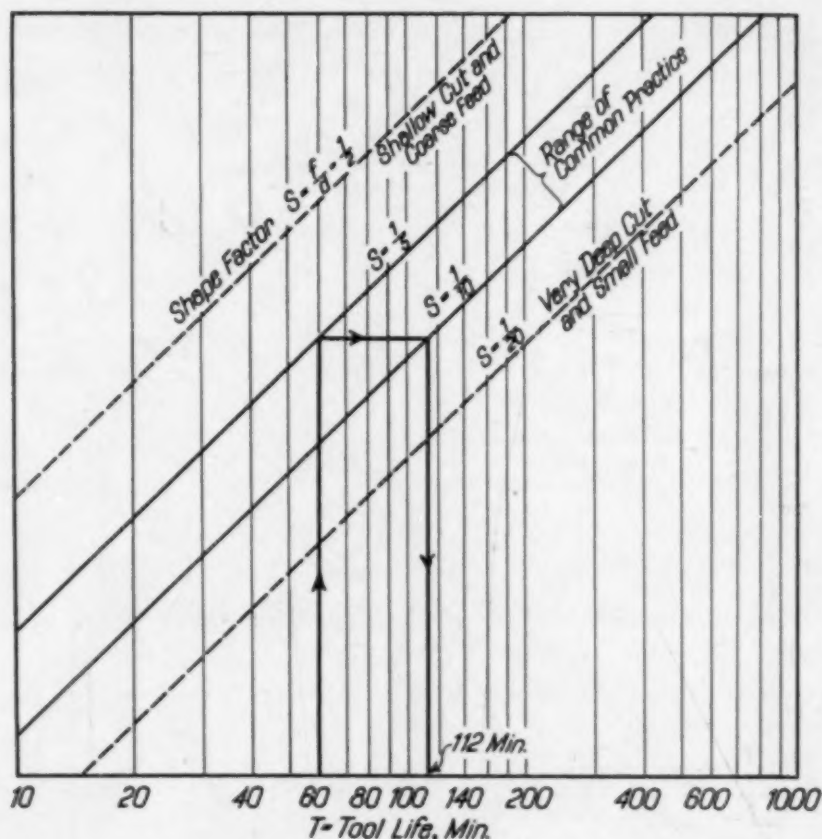


Fig. 3—Effect of Change in Chip-Shape on Tool Life.

The effect of a change in the shape factor on the tool life, for a given area and cutting speed, may be found by combining Formula 2 with Formula 3,

$$\text{whence: } T_a = T_b \cdot \left(\frac{S_b}{S_a} \right)^{\frac{(y-x)}{2n}}$$

Where T_a and T_b respectively represent the tool life for shapes S_a and S_b .

Fig. 3 is a graphical representation of this formula and shows the effect of change in the shape of chip area on endurance. The arrows refer to an example viz.: if the tool life is 60 minutes for a

shape factor $S = \frac{1}{5}$, what will be the endurance if the shape factor is changed to $S = \frac{1}{10}$? The answer, as indicated, is 112 minutes. Furthermore, it will be seen in general that tool life increases slightly if a deeper cut and a small feed is used, and decreases slightly in the case of a shallow cut and a coarse feed.

Before discussing the fourth relationship of the cutting speed (viz., as a function of the horsepower at the cutting edge) it is necessary to consider the cutting force.

Table IV
 C_p Values for Steel

True rake angle degrees	Material				
	S.A.E. 1015	S.A.E. 1025	S.A.E. 1035	S.A.E. 1045	S.A.E. 1060
35	270	312	358	406	473
30	284	328	375	426	496
25	296	342	391	445	516
20	308	357	408	464	538
15	320	370	424	481	560
10	330	383	439	498	579

Cutting force a function of chip cross section—As in the case of cutting speeds, values for the cutting force have been determined, but here again the limits are wide. Careful tests (3), however, show that the main variable upon which the cutting force depends, is again the chip cross section area A . It has been found, that the specific pressure as well as the total force give straight lines on a bilogarithmical paper as functions of the chip area. Thus it is possible to derive equations which can be used easily. The formula for the specific cutting pressure is:

$$p = \frac{C_p \cdot 1000}{E_p \sqrt{1000 \cdot A}} \text{ pounds per square inch}$$

where: C_p = a constant depending on the material to be cut and the rake angle of tool.

E_p = exponent depending upon the material to be cut.

A = area of chip section square inch.

p = specific cutting pressure pounds per square inch.

Cutting pressure is not only influenced by the material to be cut and the chip cross section, but also by the rake angle, which is taken into account in determining C_p . In addition the sharpness and finish of the tool face, the use of cutting fluids and other minor factors affect the cutting force to an extent which is not yet fully known. Due to the fact that the cutting force oscillates, the formula gives average values between the highest and lowest points of the oscillating cutting

force curve. It has been found (5) that C_p is a function of the square root of the product of the tensile strength of the material to be cut, and the lip angle of the tool, within the limits of angle commonly used. Cutting force values obtained by different investigators only show satisfactory agreement if both the tensile strength and the lip angles are taken into consideration.

Table V
 C_p Values for Cast-Iron

True rake angle degrees	Brinell Hardness					
	100	120	140	160	180	200
35	92	98	105	111	116	121
30	97	104	112	118	124	128
25	103	110	118	125	131	136
20	108	116	125	132	138	143
15	113	122	131	139	144	150
10	119	127	137	145	151	157

Table VI
 C_p Values for Other Materials

Material	True Rake Angle	Brinell	C_p
Cast steel	18°	135-150	268
Light alloy	40°	53	35
Brass	21°	Composition: 85% Cu, 9% Sn, 6% Zn	129

Tables IV and V give the C_p values for steel and cast iron. Table VI gives the values of C_p for some other materials. The true rake angle in these tables is understood to be the rake angle in the direction of the flow of the chip, (i.e., approximately perpendicular to the main cutting edge.)

The *cutting force* F follows the law:

$$F = C_p (1000 A)^r \quad (5)$$

where:

$$r = 1 - \frac{1}{E_p}$$

At first sight formula (5) appears to be too complicated for use in practice, but in reality it is very simple to calculate the cutting force if the multiplication factors from Table VII are used.

Thus, for calculating the cutting force it is only necessary to multiply the value of C_p taken from Tables IV, V or VI by the value from Table VII.

Example: Required the cutting force for S.A.E. 1035, true rake angle 20 degrees, chip cross section $A = 0.01$ square inch.

Procedure: From Table IV: $C_p = 408$

From Table VII: Multiplication factor = 6.4

Therefore cutting force

$$F = 408 \times 6.4 = 2610 \text{ pounds}$$

It will be observed from Table V that the cutting force does not increase in proportion to chip area, but at a lower rate. For example, the cutting force for $A = 0.04$ square inch and under the same conditions as given above, amounts to: $408 \times 19.3 = 7875$ pounds. This is approximately three times the value of the cutting force of 2610

Table VII
Multiplication Factors for Calculation of Cutting Force

A: Area of Chip in ²	(1000A) ^r Multiply "C _p " (Tables IV, V, VI) for the Following Metals by:				
	Steel	Cast Iron	Cast Steel	Brass	Light Alloy
0.001	1.00	1.00	1.00	1.00	1.00
0.002	1.75	1.80	1.80	1.68	1.92
0.003	2.4	2.6	2.55	2.28	2.82
0.004	3.1	3.3	3.25	2.83	3.70
0.005	3.6	4.0	3.9	3.35	4.55
0.008	5.3	6.0	5.9	4.8	7.1
0.010	6.4	7.3	7.1	5.6	8.7
0.015	8.8	10.4	10.0	7.6	12.8
0.020	11.2	13.3	12.7	9.4	16.6
0.025	13.3	16.0	15.4	11.2	20.5
0.030	15.2	19.0	18.0	12.8	24.5
0.040	19.3	24.4	23.0	15.7	32.0

pounds although the area of chip cross section has been quadrupled from $A = 0.010$ square inch to $A = 0.04$ square inch.

Cutting pressure as a function of depth of cut and feed per revolution—As in the case of the cutting speed, it is sometimes held that it does not suffice to specify the cutting force as a function of the cross sectional area of the chip, but that allowance must be made also for its shape. Under this consideration the cutting force law is often presented in the following general form:

$$F = C d^x f^y \quad (\text{Depth of cut and feed formula}) \quad (6)$$

where: F = Cutting force (lbs.)

C = Constant of material to be cut

d = Depth of cut (inches)

f = Feed (per rev.) (inches)

x, y = Exponents

Such a formula can, however, always be transformed into:

$$F = C \cdot A^{\frac{1}{2}(y+x)} S^{\frac{1}{2}(y-x)} \quad (\text{area and shape formula}) \quad (7)$$

where: A = Area of chip = $f \cdot d$

S = Shape factor of chip = $\frac{f}{d}$

It indicates—as in the case of the cutting speed—that the chip cross section A is of a considerably greater effect on cutting pressure than the shape of the chip.

Taking as an example the cutting force formula for turning steel, which was developed by F. W. Taylor, we have:

$$F = C \cdot d \cdot f^{0.934}$$

which, by the above transformation becomes

$$F = C \cdot A^{0.967} \cdot S^{-0.033}$$

The effect of a change in the shape-factor S on the cutting force is shown in Fig. 4. It will be seen that such a big change as 1:10

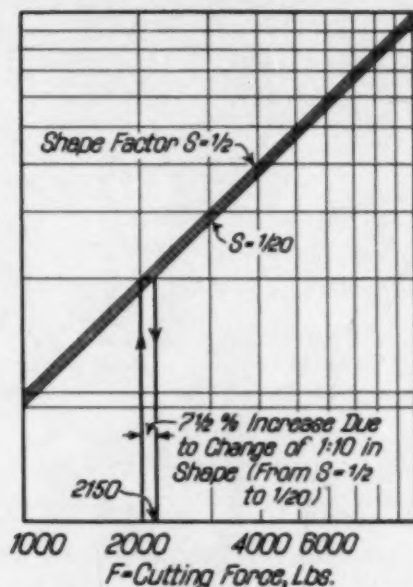


Fig. 4—Effect of Change in Chip-Shape on Cutting Force.

viz., from $S = 1/2$ to $S = 1/20$ results in an increase of the cutting force of only $7\frac{1}{2}$ per cent. A change of the area of chip of 1:10 results, however, in 830 per cent increase of the cutting force.

As will be evident from the comparison of 830 per cent change in cutting force as against $7\frac{1}{2}$ per cent, the influence of the cross sectional *shape* has been considerably overrated.

With the development of devices for measuring the cutting pressure since Taylor, it was found that the effect of the shape of the chip is practically without importance (3).

Cutting speed as a function of horsepower—The turning time t for a given piece is in general:

$$t = \frac{L \cdot z}{f \cdot n} = \frac{L \cdot z \cdot d \cdot D \pi}{12 \cdot A \cdot v} = \frac{L \cdot D \cdot K \cdot \pi}{12 \cdot A \cdot v} \text{ minutes} \quad (8)$$

where: A = chip cross section (in.²) = $f \cdot d$
 D = diameter of piece (in.)
 d = depth of cut (in.)
 f = feed per revolution (in.)

K = stock to be removed on radius (in.)

L = turning length (in.)

n = revolutions per minute

v = cutting speed (feet per minute) = $\frac{D \cdot \pi \cdot n}{12}$

z = number of cuts = $\frac{K}{d}$

For the same work piece, i.e., where L , D , K are constant values, the time t varies with the product $A \cdot v$, which is the chip volume (cubic inches per minute) and which is dependent on the horsepower at the cutting edge of the lathe tool. The greater the available horsepower, the greater becomes the permissible value of $A \cdot v$ and the smaller the turning time t . Cubic inches per grind are obtained by multiplication of $A \cdot v$ by the tool life.

The relationship between the cutting speed v_M , the chip cross section A , the horsepower at the tool P_t , and the material constant C_p follows from the fundamental law of mechanics, viz., power P equals the product of force F and velocity V :

$$P_t = \frac{F \times V}{33000} \quad \text{H P}$$

Thus:
in our case $v_M = \frac{33000 P_t}{F} = \frac{33000 P_t}{C_p \cdot (1000 A)^r}$ feet per minute (9)

This may be described as the second cutting speed law, or "machine-law".

The basic relationships in machining with single point tools—By comparison of formula (9) with formula (1) (the first cutting speed law—or tool law) the very important fact will be seen, that two different relationships exist between the chip cross section A and the cutting speed V . Basic principles which govern machining by single point tools can be derived from this fact as indicated graphically by means of the chart, Fig. 5.

In this chart the co-ordinates are the area of the chip cross section A and the cutting speed V . Their relationship is represented by two series of straight lines. The lines with the smaller slope are drawn according to the first cutting speed law (formula 1); and they may be called "*tool lines*" since they include the material of the tool due to the C_v values (Table I). The lines with the greater slope correspond to the second cutting speed law (formula 9); and may be called "*machine lines*" since their position on the diagram depends upon the horsepower of the machine.

The method of using this chart to explore the interrelated variables shown, and to determine their most effective combination is illustrated by the following example.

Assuming a machine of 8 horse power net, a high speed steel tool 18-4-2 which shall stand 60 minutes, and a chip cross section of A

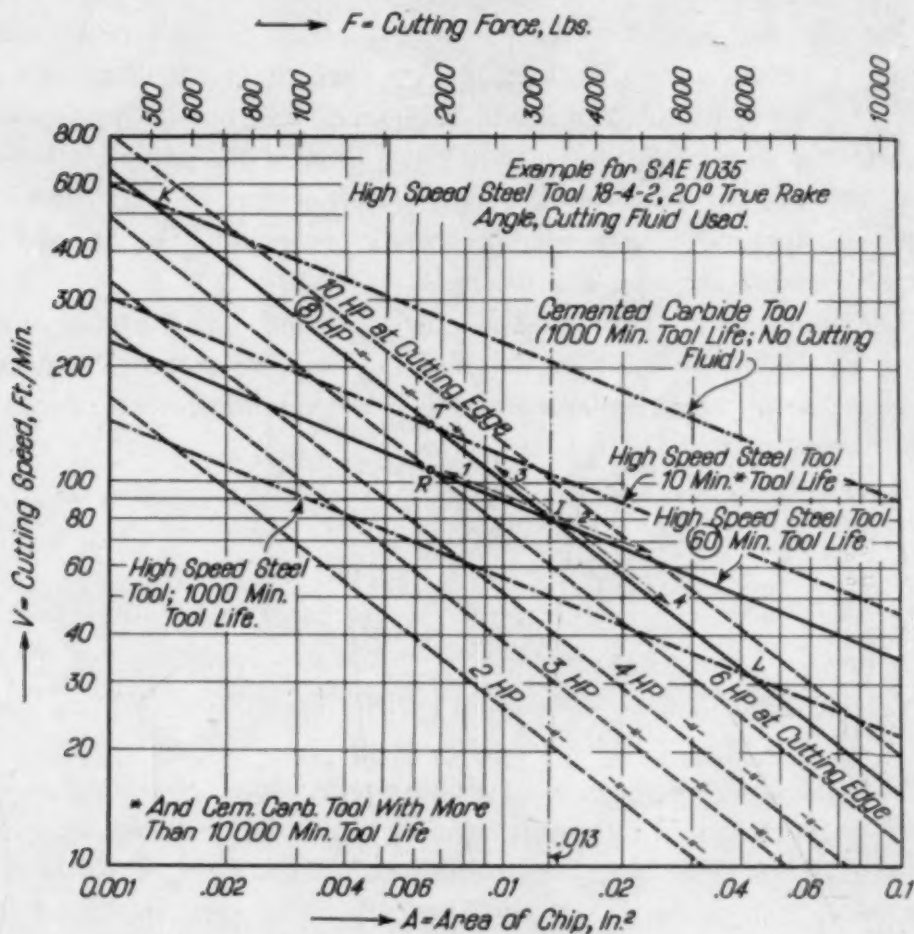


Fig. 5—Determination of the Most Effective Combination of Cutting Variables. (Productivity Diagram).

= 0.013 square inch, then it will be seen from the intersection point I, that the corresponding cutting speed would be 80 feet per minute. This corresponds to the value obtained by calculation in the example given on page 729. It is now desired to investigate the changes in the permissible cutting speed if the area of chip is changed.

1. Following the "tool line" in the direction of the arrow 1 results in an increase of the cutting speed V and a decrease of the area of chip A . In this manner, however, the "machine line" for 6 H.P. is approached. This means that a machine of 8 H.P. cannot be used to its full capacity if a smaller area than that which

corresponds to point I (and therefore to the cutting speed formula 1) is chosen, and if a tool life of 60 minutes must be maintained. Therefore the productivity would decrease considerably as will be seen from the chip volume per minute $A \cdot v$, which decreases from

$$0.013 \times 80 \times 12 = 12.5 \text{ cubic inches per minute (point I) to} \\ 0.0065 \times 105 \times 12 = 8.2 \text{ cubic inches per minute (point R).}$$

Therefore: Increasing the cutting speed by decreasing the chip cross section in such a manner that the kind of tool and the tool life is kept constant, is not advisable, due to fact that the productivity is decreased

2. Following the tool line in the direction of arrow 2 results in an increase in the chip section and a decrease in the cutting speed for a constant tool and tool life.

Here the 10 horse power line is approached, therefore such a change would result in an *overloading of the machine*.

3. Another possibility, viz., following the "machine line" in the direction of arrow 3 results in an increase of the cutting speed V , and a decrease of the area A , but here we leave the "tool line" of 60 minutes endurance, and approach the 10 minutes tool line. Since such a short tool life would be uneconomical, the kind of tool must be changed. Thus, a cemented carbide tool must be used if the cutting speed is increased "along" the machine-line of 8 horse power between the points I and K. Beyond point K the life of a cemented carbide tool would be less than 1000 minutes, while it is obviously much higher than this in the neighborhood of point I.

An additional point may be brought out here, as to the chip volume and the cutting force. Diminishing the area of chip from 0.013 (point I) to 0.0065 square inch (point T) and increasing the cutting speed from 80 feet per minute (point I) to 144 feet per minute (point T) results in a decrease of the chip volume per minute from 12.5 cubic inches per minute (point I) to 11.2 cubic inches per minute (point T).

This behavior is due to the aforesaid increase of the specific cutting pressure for decreasing chip cross sections. The cutting force, however, decreases from (see scale at top of Fig. 5) 3200 pounds at point I to 1800 pounds at point T.

This decrease in the cutting force results in a decrease of the load which is exerted on the work piece, and consequently, in less deflection and higher accuracy.

In spite of the diminution of the chip volume for increasing

cutting speeds, it is therefore often desirable to use small feeds (or small areas) and high cutting speeds. This will obviously be at some sacrifice in productivity, but still considerably more advantageous than in the case of following the direction of arrow 1, where the chip volume was reduced to 8.2 cubic inches per minute for the same area of chip $A = 0.0065$ square inch.

In addition, the use of high speeds is important for obtaining a smooth finish on the work piece due to the change in the behavior of the "built-up edge" with increasing speeds (17).

4. Following the "*machine line*" in the direction of arrow 4 results in a decrease of the cutting speed and an increase of the area of chip. Here the tool life would be improved by approaching the 1000 minute tool line, and would also increase the chip volume. At point L this would be $0.041 \times 31 \times 12 = 15.3$ cubic inches per minute. Thus it seems that the use of a low cutting speed in connection with a big chip results in both a better endurance for the same kind of tool and a higher productivity. However, it must be borne in mind that the cutting force increases considerably in this case, for instance from 3200 pounds (point I) to 8000 pounds (point L). The work piece and the lathe bed would therefore be deflected in a considerably higher degree than in the case of point I, moreover the feed per revolution would be coarse; both resulting in a lower accuracy and poorer finish. This procedure (along arrow 4) is only advisable if roughing cuts are to be taken on stable work pieces.

Therefore: Selection of the cutting speed or changing it as a function of chip area, should be made in accordance with the "*machine line*", and the tool should be adapted to the respective speed. Increasing the speed should often be accomplished by the use of a higher quality tool and is advisable if the work piece is unstable and if higher finish is desired. It is advisable on the other hand to decrease the speed and increase the area of chip in the case of roughing big work pieces.

Effect of chip shape on productivity—Since the machine line should be used for selecting the cutting speed, and the tool should be adapted to this speed, it is necessary to consider the effect of a splitting up of the tool line according to different shape factors.

Reference is made to Fig. 6 which is partly a repetition of Fig. 5. The tool life line for 60 minutes endurance is split up here in 4 lines according to shape-factors 1:2, 1:5, 1:10 and 1:20, while the other tool life and horsepower lines of Fig. 5 are omitted.

Taking again point I as a starting point, it is seen that by tracing along arrows 5 and 6 respectively, the cutting speed for $A = .013$ square inch varies between 100 feet per minute (point N) and 71 feet per minute (point O). At the same time, however, the horsepower needed at the cutting edge (in the case of point N) would be greater than 8 horsepower, i.e., the machine would be overloaded, while it would not be utilized fully in the case of point O.

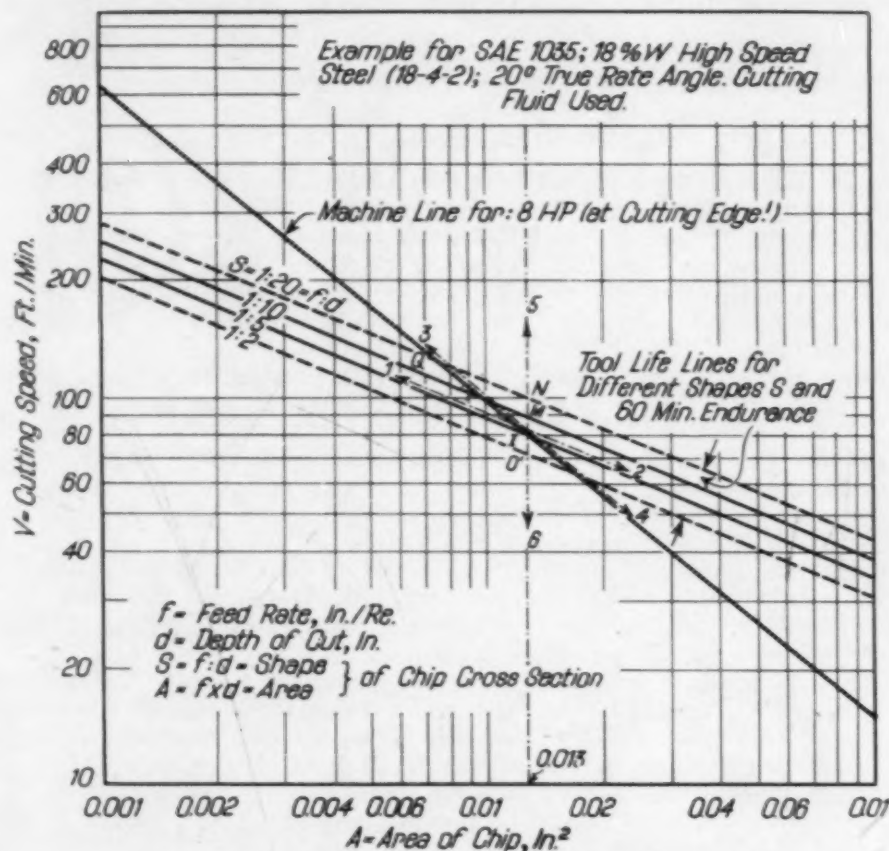


Fig. 6—Effect of Chip-Shape on Utilization of Machine.

It is therefore not possible for a given chip cross section to correlate a change of the chip shape with the proper utilization of the machine by increasing (or decreasing) the cutting speed as against the tool life line.

If, however, the cutting speed is kept at 80 feet (point I) but a shape of 1:10 is taken, then the tool life will increase as mentioned above. It seems to be fair to leave this "margin of tool life" within the values C_v which is mostly in favor of the user of Table I as he will obtain a larger tool life than 60 minutes for values of S smaller than 1:5. On the other hand, it would not be advisable to

lower the C_v values down to a shape of $S = 1:2$ as this ratio occurs only occasionally in practice.

In any event, as shown by Fig. 6, the effect of a change in shape within the limits normally used, is relatively small.

From the foregoing it can also be derived that future development will follow the same trend as is found in all branches of machinery, viz., the utilization of power by high speeds and low forces. This results, as far as the tools are concerned, in the use of high quality tools, and for the machines, in high powered machines of rigid construction and maximum freedom from vibration.

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MAGNETIC PERMEABILITY OF SOME AUSTENITIC IRON-CHROMIUM-NICKEL ALLOYS AS INFLUENCED BY HEAT TREATMENT AND COLD WORK

By J. B. AUSTIN AND D. S. MILLER

Abstract

The 18-8, 18-12 and 25-12 type alloys, if free from ferrite, have substantially the same permeability, namely, 1.003 (air = 1.000), which is low enough to permit the use of the alloys in applications for which ferrous materials have hitherto been unsuited, as for example, in construction near the compass of a ship. If ferrite is present, whether it be high temperature (delta) ferrite, ferrite formed by cold work, or ferrite resulting from carbide precipitation, the permeability of the alloy is increased.

The formation of ferrite by cold work, hence, the increase in permeability, is most marked in 18-8, is considerably less in 18-12 and is negligible in 25-12 type except for the most severe working. If minimum permeability is desired the use of the 18-12 alloy rather than the 18-8 gives an added margin of insurance against an increase in permeability due to improper handling. The permeability of material of all three types, when free from ferrite, is well below the maximum of 1.02 permitted for such material in the specifications of the United States Navy Department. The permeability of 18-8 which has been cold-worked may exceed the maximum of 1.05 permitted for hardened material in the Navy specifications, but the permeability of 18-12 or 25-12 is not likely to reach this value unless the alloys have been very severely cold-worked. Careful welding does not increase the permeability of these alloys.

THE fact that the austenitic iron-chromium-nickel alloys combine great strength with low magnetic permeability has led to their use in many applications in which ferrous materials have hitherto not been employed, as for example, in armature binding wire and as a "nonmagnetic" material for use near the compass of a ship. As the

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available data on the magnetic properties of these alloys are meager, measurements have been made of the permeability of the more important commercial alloys of this type, particularly as regards the influence of cold work and welding. Inasmuch as the appearance of ferromagnetism in an austenitic alloy is a very sensitive indicator of the presence of ferrite, these observations also have a bearing on the structural changes occurring in such alloys during mechanical treatment and on the effectiveness of some heat treating operations in removing ferrite. The results show (1) that the 18-8, 18-12, and 25-12 alloys when free from ferrite have substantially the same permeability, which is very close to 1.003; (2) that the permeability is increased by cold work, the change being marked in 18-8, considerably less so in 18-12, and almost negligible in 25-12; (3) that careful welding does not appreciably increase the permeability of these alloys, although if a minimum permeability is desired the use of 18-12 welding rod provides an added margin of security.

METHODS AND APPARATUS

The chief method of investigation consisted in weighing the force exerted by a nonuniform magnetic field on a sample of known mass. This method is strictly applicable only to paramagnetic (i.e., very feebly magnetic) substances whose permeability is sensibly independent of the magnetizing force; but experience has shown that reasonable accuracy can be obtained in the measurement of permeabilities as high as 1.2. Beyond this point the errors become increasingly larger and the results less accurate. The apparatus (Fig. 1), originally designed for another purpose and later adapted to the requirements of these measurements, has been described in detail elsewhere.¹ It suffices to state that it consists of a sensitive analytical balance from one pan of which there hangs a glass fiber which suspends the sample at the position of bucket B (Fig. 1) so that it is on the axis of a solenoid producing the nonuniform field. In most of the measurements the specimens were small pieces of strip or wire about 2 centimeters in length, weighing approximately 1 gram, and in such cases the force measured varied with the sample from 1 to 50 milligrams. In the examination of welds, specimens weighing 50 grams or more were used and with these the force varied from 100 to 500 milligrams. On account of the wide variation in the magni-

¹Sosman and Austin, *Journal, Washington Academy Sciences*, Vol. 25, 1935, p. 15.

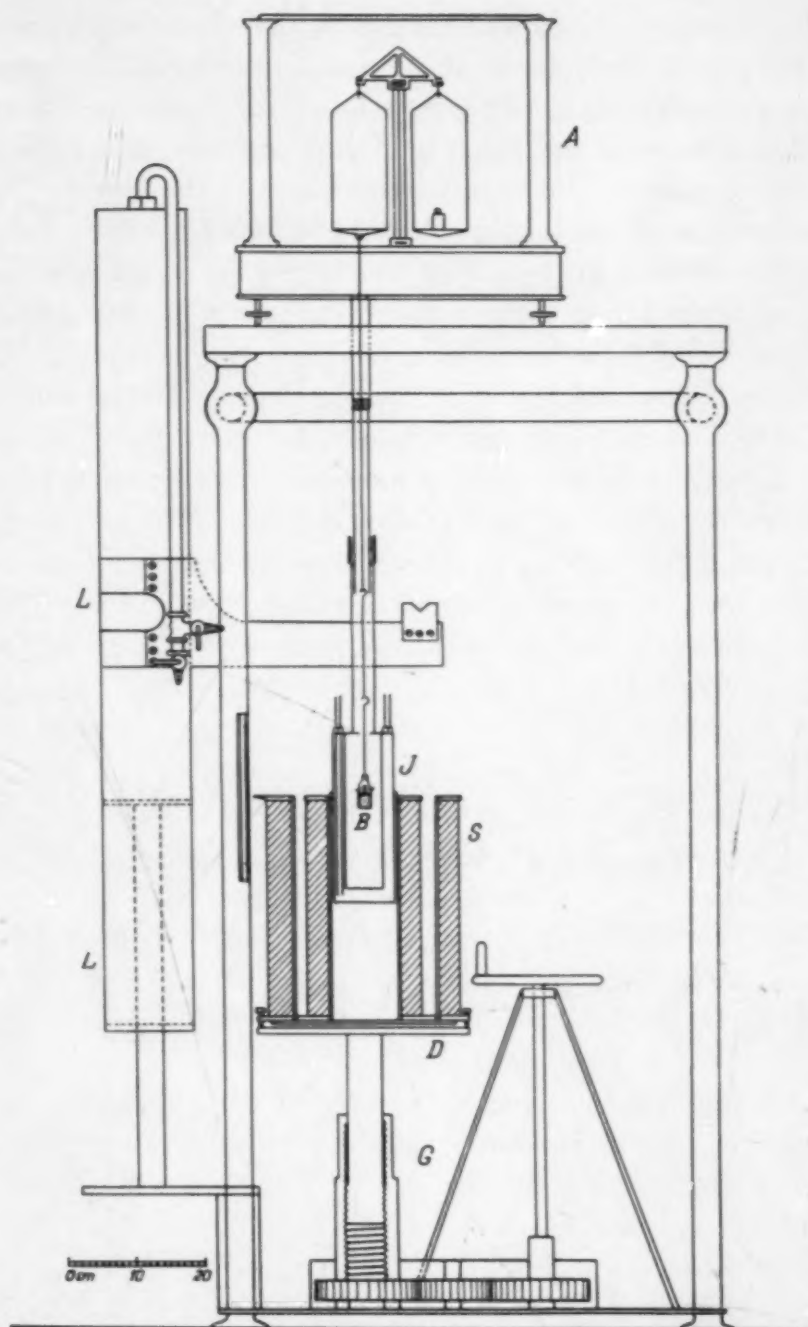


Fig. 1—Diagrammatic Sketch of Apparatus Used in Determining Permeability by the Weighing Method.

tude of this force a statement of accuracy in terms of permeability is misleading; a more informative statement is that the detectable variation is about 0.05 milligram. The magnetizing force used was in every case 436 oersteds, which appears to be sufficient to produce magnetic saturation in those samples in which a small quantity of

ferrite is present. In view of the fact that for these samples the permeability is a function of the magnetizing force, determinations using a value other than 436 oersteds will yield slightly different results, and a magnetizing force greater than this value will give a lower permeability. Unless otherwise stated, the direction of the field was parallel to the long axis of the samples, which was also the direction of drawing in the cold-worked samples. This method gives a direct measure of the mass susceptibility (X). To convert this quantity to permeability it is necessary first to calculate the volume susceptibility K by multiplying X by the density of the sample; the permeability μ , is then obtained through the relation $\mu = 1 + 4\pi K$. The densities used in the calculations were taken from previous determinations on alloys of similar composition. Unless the contrary is specifically stated all data were obtained by this weighing method.

A few measurements have also been made by a ballistic method which is particularly useful for the range of permeability above 1.2 in which the accuracy of the weighing method decreases. The apparatus used was essentially similar to that commonly used in obtaining the magnetization curves of a ferromagnetic material. The field is produced by a small electromagnet whose current is supplied by a storage battery to minimize fluctuations. Between the poles of the magnet is a search coil of 3700 turns so mounted that a cylindrical specimen $\frac{1}{4}$ inch in diameter may be inserted through holes in the center of the pole faces. A ballistic galvanometer is connected to the search coil. For a field of about 50 oersteds the galvanometer deflection is between 150 and 200 scale divisions and as, under favorable circumstances, readings can be made to the nearest fifth of a scale division, it is possible in such cases to determine a permeability near unity with an accuracy approaching 0.1 per cent. The results by this method agree within the error of measurement with those obtained by the weighing method; for example, in one instance the permeability was 1.0122 as determined by the weighing method and 1.0135 by the ballistic method.

SAMPLES

The chemical composition of the samples tested is given in Table I. All samples were etched before measurement to remove any ferromagnetic material formed during the preparation of the sample, and in some cases check measurements were made after a second

etching to make sure that all effects of cutting the specimen had been eliminated. The amount of metal removed from the surface by the double etching was of the order of 0.001 inch. The magnitude of the error likely to be introduced by failure to remove ferrite formed

Table I
Chemical Composition of Samples

Type	Sample No.	C	Mn	P	S	Si	Cr	Ni	
18-8	60A	0.08	0.30	18.15	8.70	
	18-8W	0.07	0.34	0.008	0.028	0.34	17.56	9.27	
	229A	0.08	0.40	0.014	0.008	0.24	18.13	8.94	
	191A								
18-8 Ti	5203								
	5204	0.07	0.59	0.013	0.003	0.47	18.08	9.12	Ti 0.38
	5205								
	5210								
18-8 (High Carbon)	18-8 C	0.15	0.42	0.030	0.009	0.48	16.63	8.31	
18-8 Se	18-8Se	0.09	0.99	0.147	0.050	0.35	18.88	9.32	Se 0.20
18-12	171A	0.08	0.50	0.10	0.014	0.29	17.04	12.67	
	173A	0.08	0.57	0.018	0.009	0.21	16.47	11.32	
	18-12W	0.09	0.50	0.011	0.016	0.18	17.72	10.62	
25-12	25-12W	0.09	1.92	0.010	0.082	0.41	20.51	11.87	

during preparation is illustrated in Table II which further shows that particular care must be used as the permeability approaches unity.

Table II
Effect of Pickling on the Apparent Permeability of Small Specimens of a Sample of 18-8 (229A)

Per Cent Reduction in Cross Section by Cold Rolling	Permeability (μ)	
	After Light Etch	After Second Etch (Total of 0.001 Inch from Surface)
0	1.0027	1.0027
2.4	1.12	1.05
4.8	1.13	1.05
9.1	1.20	1.10
14.9	1.35	1.32

RESULTS AND DISCUSSION

The results for a number of samples are given in Tables III to VIII. The data in Table III, which summarize the measurements on samples carefully heat treated to remove ferrite, show clearly that when properly heat treated all of the materials tested are "nonmagnetic," and that all have a permeability close to 1.003 which is somewhat lower than that given for 18-8 by the Joint Research Committee² or the permeability indicated by the recent measurements of Buehl,

²Proceedings, American Society for Testing Materials, Vol. 32, Pt. I, 1932, p. 167.

Hollomon and Wulff.³ The cause of this difference is not clear. It may arise in part from the difference in method, the ballistic method being employed by the Joint Committee, and a torsion method by Buehl, Hollomon and Wulff, but it seems more likely that it is

Table III
Permeability of Stainless Alloys, After Heat Treatment to Remove Ferrite

Type	Sample No.	Permeability at 436 Oersteds
18-8	229A	1.0027
	18-8W	1.0029
18-8 (high carbon)	18-8C	1.0034
18-8 Se	18-8Se	1.0030*
Stabilized 18-8 Ti	191A	1.0037
18-12	173A	1.0028
	18-12W	1.0033
25-12	25-12W	1.0031

*Ballistic method.

due to differences in heat treatment of the samples studied in spite of the fact that the treatment to which the alloys are stated to have been subjected should have been sufficient to render the alloys free from ferrite. Taking into account the relatively greater accuracy of the weighing method at low permeabilities, and considering the fact that the results of the other investigations are not in close agreement while ours are reproducible, it is believed that the values given in Table III for properly heat treated alloys are more accurate than the others mentioned. It is not claimed that these values are final since a number of factors affecting permeability, the distribution of the magnetic phase, for example, have not been given much attention, but it seems safe to say that further investigation will not appreciably alter the values given. Moreover, while only the more important alloys were studied it can be stated with some confidence that the permeability of all the austenitic chromium-nickel steels, if free from ferrite, will be close to 1.003. It is interesting to note that the value 1.003 corresponds closely to the permeability of pure iron in the gamma form above 900 degrees Cent. (1650 degrees Fahr.)

EFFECT OF COLD WORK

Although the permeability of all these alloys, in the ferrite-free condition, is the same, it differs markedly after cold work, as is evident from Table IV. Both 18-8 and 18-12 develop ferromagnetism

³Buehl, Hollomon and Wulff, *Metals Technology*, Oct. 1939 T.P. 1120.

under such treatment, but the effect is much greater in the 18-8; thus, the permeability of a sample of 18-8 strip after 2.4 per cent reduction of thickness was greater than that of an 18-12 wire after 50 per cent reduction of area. The permeability of the 25-12 alloy is practically unchanged after 90 per cent reduction. The slight decrease in

Table IV
Effect of Cold Work on Permeability (at 436 Oersteds) of Stainless Alloys

Per Cent Reduction of Cross Section by Cold Rolling or Drawing	18-8		18-12		25-12
	229A Strip	18-8W Wire	173A Wire	18-12W Wire	25-12W Wire
0	1.0027	1.0029	1.0028	1.0033	1.0031
2.4	1.050
4.8	1.052
9.1	1.100
10.	1.10	1.0065	1.0030
15.	1.32
20.	1.4	1.019	1.0029
30.	1.035	1.0027
40.	1.071	1.0031
50.	1.0185	1.11	1.0034
60.	1.20	1.0037
70.	1.3	1.0041
80.	1.4	1.005
85.	2.0	1.5	1.005
90.	1.7	1.005

permeability of 25-12 for small reductions, which is encountered in other metals, is probably the normal effect of cold work and would be observed in the other group were ferrite formed less easily. Since 18-12 and 25-12 are much less sensitive to cold work than 18-8 they offer considerable advantages over it if a minimum permeability is desired in metal which is to be subjected to much working. The recently reported data of Horwedel⁴ are, as nearly as one can read from the diagram in which they are presented, in satisfactory agreement with the data in Table IV.

We have also calculated the induction (B-H) for sample 229A from our data and have compared it in Fig. 2 with data on the same material given by Buehl, Hollomon and Wulff, as read from the diagrams in their paper. It is evident that the agreement is satisfactory. It should be noted that these values for the magnetic induction are also consistent with the measurements of Aborn and Bain,⁵ who determined the induction of samples containing a different amount of ferrite in studying phase relationships in these alloys.

⁴Horwedel, Thesis, Ohio State University, 1935 reported by Kinzel and Franks, *The Alloys of Iron and Chromium*, Vol. II, Engineering Foundation. McGraw-Hill, New York, 1940, p. 339.

⁵Aborn and Bain, *TRANSACTIONS, American Society for Steel Treating*, Vol. 18, 1930, p. 837.

The sensitivity to cold work is related to the structural stability of the alloy, specifically, to its ability to preserve austenite. Austenitic alloys containing approximately 18 per cent chromium and 8 to 12 per cent nickel normally encountered in practice are not in their equilibrium condition, which would be largely ferritic at ordinary temper-

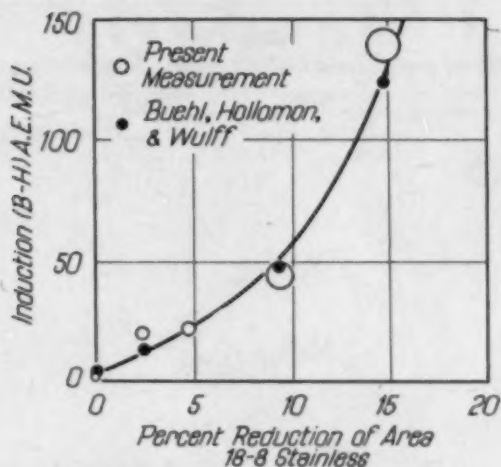


Fig. 2—Comparison of Magnetic Induction (B-H), of Samples of Cold-Reduced 18-8 with Measurements by Buehl, Hollomon and Wulff on the Same Material.

ature, but they are prevented from changing to ferrite by the substantially zero rate of spontaneous transformation at these temperatures. A detailed account of the phase relationships in these alloys, and in particular of the effect of cold work, is given by Aborn and Bain (Reference 5). Cold work, however, acts as an accelerator and tends to produce ferrite, but the magnitude of the effect accompanying a given amount of cold work depends largely on the ratio of the concentration of chromium to that of nickel and of carbon, both of which tend to stabilize austenite. In addition, nickel appears to contribute slightly to the sluggishness of the transformation. For a given chromium content an 18-8 type alloy which is high in carbon or in nickel or in both, is altered to a lesser degree by a given amount of cold work than a similar alloy low in carbon or in nickel. Thus, it has long been realized that the 18-12 type alloy is much less sensitive than the 18-8 type, but it does not appear to be widely recognized that small changes in the nickel-chromium ratio within the nominal tolerance of the latter type are also significant. That this ratio is a factor to be considered can be seen by comparing the permeability of samples cold-worked to the same extent, having approximately the

same carbon content but differing a little in the relative amounts of chromium and nickel. Such a comparison has been made in Fig. 3 which contains two curves, one for three alloys which had been reduced 10 per cent and another for two alloys which had been reduced 50 per cent in cross section. A third point is available for the latter

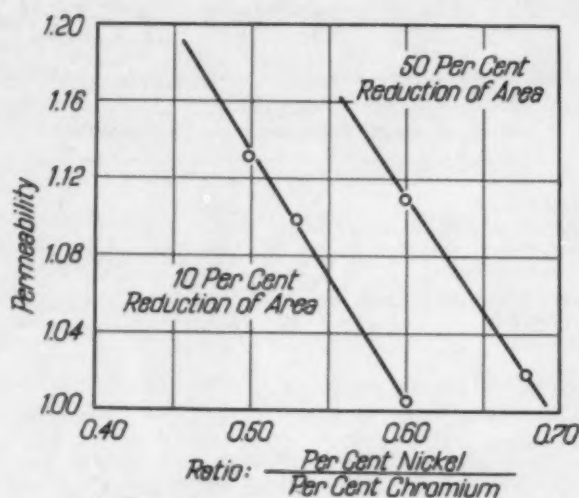


Fig. 3—Variation with Chemical Composition of Permeability after a Given Amount of Cold Reduction.

curve but falls far above the top of the diagram. The steepness of these curves is rather striking, and the conclusion to be drawn from the diagram is that in all cases where minimum permeability is an important factor in an alloy of the 18-8 type the nickel-chromium ratio should be kept as high as is compatible with other requirements. In this connection it should be noted that if minimum permeability is desired, care should be used in handling 18-8 material which has been treated to obtain a very low permeability since rough treatment such as hammering or bending may cold work the material locally enough to produce spots of ferrite. Thus, a piece of 18-8 which is acceptable when tested may, if abused in handling, become unacceptable by the time it is put into service.

Table V shows that nearly all the ferrite formed in 18-8 by several passes of cold drawing is removed by a 15-minute treatment at 1050 degrees Cent. (1920 degrees Fahr.); in fact, our experience with cold-worked alloys indicates that this period is more than enough at this temperature, unless the absolute minimum permeability is desired. At 1150 degrees Cent. (2100 degrees Fahr.) 4 or 5 minutes is ample time for the complete annealing of small specimens.

It was also of interest to investigate the permeability of these

alloys in different directions. Experience with other materials indicates that a paramagnetic substance possessing cubic symmetry should be magnetically isotropic, that is, the permeability should be the same in all directions, but that as a ferromagnetic phase develops directional effects may appear. The behavior of the stainless alloys is in entire agreement with this generalization. Four typical samples were

Table V
Effect of Heat Treatment on Permeability

Alloy	Treatment	Permeability
18-8W	As-received	1.015
	Quenched after 5 mins. at 1150° C. (2100° F.).....	1.0029
Stabilized 18-8 wire (191A)	Cold drawn followed by 24 hrs. at 870° C. (1600° F.) Fur- nace cooled	1.0038
	Cold drawn followed by 16 hrs. at 870° C. (1600° F.), quenched	1.0037
Stabilized* 18-8 tube		
5203	Hot-rolled (as received)	1.053
5204	Hot-rolled, 15 mins. at 1050° C. (1920° F.) quenched in water	1.018
5210	Hot-rolled, heated to 1050° C. (1920° F.) quenched in water; cold drawn 2 passes, no intermediate anneal; again heated to 1050° C. (1920° F.) and water quenched; cold drawn 3 passes no intermediate anneal; heated to 1050° C. (1920° F.) quenched in water	1.015
5205	Hot-rolled, ½ hr., at 1200° C. (2200° F.) quenched in water	1.009
Stabilized 18-8 bar		
334A (191A)	Hot-rolled followed by 2 hrs. at 870° C. (1600° F.)	1.0117
335A (191A)	Annealed 30 mins. at 1950° F. followed by 2 Hrs. at 870° C. (1600° F.)	1.0064
18-12 wire (173A)	15 mins. at 1150° C. (2100° F.)—Quenched	1.0028

*Examination under the microscope shows that 5203 contained about 1 per cent of high temperature ferrite. This was diminished to 0.1 per cent in 5204, to a trace of 5205, and to a point below certain identification in 5210.

selected and their permeability was measured both parallel to, and perpendicular to, the direction of drawing. Since measurements in the latter direction required that the specimen be suspended perpendicular to the axis of the solenoid some uncertainty as to the magnetizing force was introduced but the magnitude of this effect could be estimated by using specimens of two sizes. The results are shown in Table VI Cold work alone does not produce a magnetic directional effect but, as is evident from the data, cold drawing which produces ferrite does appear to make an alloy magnetically anisotropic (see Table VI) since the permeability in the direction of drawing appears to be greater than that at right angles to this direction. It is, however, not safe to base too many conclusions on this fact since the effective shape of the specimens is different in the two cases.

Table VI
Permeability in Different Directions in Wire

Type	Degree of Cold Work	25 Mm. Sample		5 Mm. Sample	
		Field Parallel Direction of Drawing	Field Perpendicular Direction of Drawing	Field Parallel Direction of Drawing	Field Perpendicular Direction of Drawing
18-8	None (annealed)	1.0027	1.0027
18-8	10% Red. of Area	1.0985	1.082
18-12	5% Red. of Area	1.0082	1.0060	1.0092	1.0067
25-12	60% Red. of Area	1.0038	1.0036	1.0037	1.0037

Another question which has occasionally been raised is whether there is a magnetic aging in these alloys, that is, whether the permeability may increase with time. This seems unlikely on general principles and no evidence for such an effect has been obtained experimentally. Tests on two samples which had previously been studied were made after about 18 months storage with results (Table VII)

Table VII
Permeability of Alloys After 18 Months Storage

	Sample No. 1, 18-8 (0.09 C) (10% Reduction of Area by Cold Work)	Sample No. 2 18-8 (High Carbon)
Permeability in first test	1.103 \pm 0.002	1.0034 \pm 0.0002
Permeability after 18 months	1.102 \pm 0.002	1.0034 \pm 0.0002

which showed no significant change in permeability. No information is available as to the length of time which had elapsed between the manufacture of these alloys and the first tests so that it cannot be stated that no change occurs immediately after heat treatment, machining or drawing, but it can be stated from the low permeability of the samples tested that if such an effect occurs it is very small and must be confined to a relatively short interval after manufacture.

WELDS

The data for welds on $\frac{3}{8}$ -inch plate are given in Table VIII. The permeability was measured in each case before and after etching in order to correct for the ferrite formed in cutting the sample. It is evident that it is easily possible to keep the permeability low but that it may be advisable in dealing with thick plates to use 18-12 welding rod. In addition to the results shown in the table, two experimental arc welds made with $\frac{1}{32}$ -inch sheet of the 18-8 type (exact

composition unknown) were examined; the permeability in each case was 1.004.

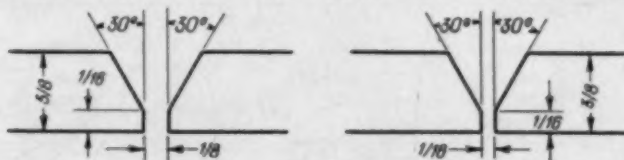
It should be emphasized that although the permeability provides a measure of the amount of ferrite formed under various conditions, it has not yet been established that the presence of low temperature or cold work ferrite in itself diminishes the corrosion resistance of this alloy in any environment to which the alloy is adapted, that is, in all cases where the oxide film is maintained. In fact there is some

Table VIII
Permeability of Stainless Alloys Steel Welds

	1	2*	3
Parent Metal	18-8 (60A)	18-8 (60A)	18-12 (171A)
Welding Rod (coated)	18-8	18-12 (173A)	18-12 (173A)
Type of Joint	No. 1 below	No. 2 below	No. 1 below
Welding Current (amps.)	50-60	50-60	50-60
Arc Voltage	28	28	28
Permeability (unetched)	1.029	1.030	1.006
Permeability (etched)	1.009	1.020	1.004

*Bad weld, much oxide.

A sample weld of 18-8 supplied by the Metal and Thermit Corporation of Weehawken, N. J., had a permeability of 1.10 and 1.07 in the unetched and etched condition respectively.



The specimens cut from each weld included about $\frac{1}{2}$ inch of parent metal on each side.

evidence to indicate that cold-worked 18-8 resists salt spray and oxidation better than the annealed alloy, presumably because of the greater continuity of the protective film. It has also been reported⁶ that 18-8 oil-still tubes which have developed carbide-induced ferrite exhibit no loss in general corrosion resistance. Moreover, there is a whole series of ferritic chromium steels which are ferromagnetic but are practically as resistant in many environments as the corresponding austenitic chromium-nickel steel. Finally, corrosion resistance appears to be enhanced by polishing, which implies the presence of ferrite in the minute surface layer of worked 18-8.

In conclusion it should be noted that when these alloys are substantially free of ferrite (annealed), their permeability is well below the maximum of 1.02 permitted by the U. S. Navy Department (Specification 47-S-20A, May 1, 1937) for such material. Properly

⁶Newell, St. Louis meeting of the A.P.I., 1931.

heat treated material can, therefore, easily meet this requirement. The Navy specification for "hardened," i.e., cold-worked, material is 1.05. The permeability of the 25-12 alloy appears to stay well below this value even after severest cold work. The permeability of 18-12 is not likely to approach it except after cold working of the order of 80 per cent reduction in area, but the permeability of the 18-8 alloys may reach 1.05 after relatively mild cold working, especially if the ratio of nickel to chromium is on the low side of the allowable tolerance. This point is emphasized, because it occasionally happens that material of this type, which is delivered in proper condition with an acceptable permeability, may develop ferritic spots in later handling if the metal is mistreated, as by severe hammering. With proper care, however, no trouble should arise from this source.

SUMMARY

1. The austenitic iron-chromium-nickel alloys, if free from ferrite, all have practically the same permeability, i.e., 1.003 (at 436 oersteds).
2. The 18-8 and 18-12 groups become ferromagnetic if sufficiently cold-worked but the effect is very much greater in the former. The permeability of 25-12 is practically unaffected by cold work. In general, the higher the carbon and the higher the ratio of nickel to chromium the smaller the effect of cold work on permeability.
3. In none of the alloys is the increase in permeability directly proportional to the amount of cold work; in fact, there appears to be no simple relation between these quantities.
4. Careful welding does not appreciably alter the permeability of these alloys. If a minimum permeability is essential it is advisable to have the nickel-chromium ratio as high as is compatible with other requirements.